T H E S I S

by

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PART I

Temper-Brittleness.

PART II

Ac_1 and Ar_1 in Special Steels

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PART II

AC_1 and AR_1 in SPECIAL STEELS.

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NOTE The Tables, Figs., and Photomicrographs referred to in the Text are contained in a separate binding, for the convenience of the reader.

PART I

TEMPER-BRITTLENESS.

INTRODUCTION

Temper-brittleness is the name which has been applied to one of the most puzzling phenomena found in special or alloy steels. It appears to have been first known in pre-war days at Krupp's Works in Essen, Germany, where it was observed that armour-plates were liable to develop brittleness in a manner which was not understood. The name of Krupp-Krankheit was consequently commonly applied to this mysterious phenomenon at that time. During the war it became a troublesome problem in connection with the greatly increased production of armaments in Great Britain, and, as a result, a good deal of research has been carried out on the problem. The earliest investigations (1,2,3,4,5,6,7,) established the fact that if a piece of steel which was susceptible to this trouble was cooled sufficiently quickly, as by quenching in water, from the tempering temperature, it did not develop brittleness and that it was only when a slow rate of cooling was used that a notable degree of brittleness appeared.

A large amount of work has been done by the method of heat-treatment of test-bars followed by mechanical testing. The most important researches in this connection are those of Greaves and Jones.[8,9] These investigators have shown very clearly that whereas the rate of cooling exercises a very pronounced effect on the notched-bar impact value of a "susceptible" steel, yet no change takes place in any other mechanical property, with the possible exception of a small

change in the elastic limit. A large number of nickel-chromium steels (steels containing both nickel and chromium have been shown to be, in many cases, particularly susceptible to temper-brittleness) were used in their investigations and of these the majority, but not all, were found to have an elastic limit in the brittle condition about 3 tons per square inch higher than the same steels in the water-quenched tough condition. When, however, moderate rates of cooling were adopted or when the quenched steels were subjected to low temperature annealing (neither of which treatments produced brittleness in the steels), the elastic limit of the two conditions was equalised. They consequently concluded that the difference in elastic limit was due to a condition of strain produced by water-quenching and not to the difference between the brittle and the tough conditions.

As a result of the work carried out by Greaves and Jones and by other investigators the effects of variation of heat-treatment on the notched-bar impact value have now been exhaustively investigated. The most important facts which have been established in these researches are as follows:-

(1) The degree of brittleness produced is roughly proportional to the rate of cooling. The slower the cooling rate the lower is the impact value obtained on a test-piece.

(2) There appears to be a temperature range of brittleness. This varies in different steels but is, as a rule, from about 500°C up to the highest tempering temperature, usually about 650°C. If a steel is quenched or quickly cooled from a tempering temperature in this range it is relatively tough, i.e., it gives a high impact value. If it is cooled slowly it is relatively brittle i.e. it gives a low impact value.

The effect of composition on the susceptibility of a steel to brittleness has also been the subject of a good deal of investigation. Steels containing nickel and steels containing chromium have been found to be susceptible in varying degree to brittleness and when both these elements are present in any quantity in the same steel, the susceptibility, as a rule, is high. In regard to other elements, Greaves and Jones (10) have found that manganese and phosphorus increase susceptibility, especially when present in any amount beyond a small percentage, that silicon and vanadium have only a slight effect and tungsten no appreciable effect, and that molybdenum, on the other hand, has a very pronounced effect in reducing susceptibility. Andrew and Green (11) have shown that in nickel steels increase of phosphorus increases proportionately the susceptibility to brittleness, and Griffiths (12) has found that a carbon steel impregnated with nitrogen (as iron nitride) shows susceptibility similar to that found in the alloy steels mentioned above.

Up to the present time the problem of temper-brittleness has been dealt with largely from what may be called the engineering standpoint. The effect of heat-treatment and of composition on the notched-bar impact value of steels has received a great deal of attention. A certain number of investigators (whose work will be dealt with in the next section below) have attempted to find a physical change accompanying the change in impact value, but, relatively to the large amount of work which has been carried out by mechanical testing, the problem has received

scanty attention from the standpoint of pure science. The conclusions of those who have published the results of physical determinations have resulted in the general acceptance of the conclusion that no physical change accompanies the change from the tough to the brittle condition. Changes in some magnetic properties are sometimes quoted as being an exception to the above general conclusion, but, as will be shown below, the evidence of changes in magnetic properties is at present very meagre.

It is consequently not surprising that no satisfactory explanation of temper-brittleness has been put forward. Indeed very few workers have felt themselves in a position to advance any explanation whatever. Examination of the microstructure has not been found, up to the present, to give any assistance in this direction. Philpot[13] showed clearly that the fracture of brittle specimens appeared to follow the crystal boundaries whereas the tough specimens showed a ragged transcrystalline fracture, but, although in some cases a corresponding difference was obtained in the microstructure, in general no difference whatever was detectible between the structures of tough and brittle specimens. Many other observers have come to the same conclusion.

In consequence of the apparent absence of any change in microstructure or in physical properties in any way commensurate with the great change indicated by the arbitrary notched-bar impact tests, the value of all or any of the modifications of

these tests as a measure of the real toughness or brittleness of a material has naturally been questioned. From the work published up to the present it is easy to conclude that a minute and almost negligible change in the condition of a steel is magnified out of all proportion by a notched-bar impact test.

It was evident that the whole subject could never be placed on a satisfactory basis until the physical changes which produce the alteration in impact value could be laid bare. In spite of the negative results published by previous workers, it was believed that definite and detectible physical changes must take place in the transition from the tough to the brittle condition and it was in order to elucidate these that the present investigation was carried out. Success in this direction would lead to the true explanation of the phenomenon and, among practical applications which might be expected to result ultimately from this work, it will be sufficient to indicate here the possibility of devising a physical method for determinining the toughness of a material without having to resort, as at present, to the fracture of a separate test-piece, which is not necessarily in the same condition as the material which is to be used in practice.

PHYSICAL DETERMINATIONS by other INVESTIGATORS

Greaves and Jones [9] have compared the specific gravity, hardness, and specific resistance of tough and brittle specimens in which tempering effects were equalised. They found that the specific gravity of water-cooled tough steel was from 0·0004 to 0·0010 less than that of the same steel in the brittle condition, but that when the tough steel was reheated to 200°C this difference was greatly reduced. This rise in specific gravity on low-temperature annealing they attributed to the removal of a condition of strain in the water-quenched steel, and they concluded that "the difference in density due solely to the condition of toughness or brittleness appears to be at the most 1 part in 20,000".

From Brinell hardness determinations on water-quenched and slowly cooled specimens they conclude: "These results show that when unequal tempering effects are eliminated in producing tough and brittle material, there is no appreciable difference in the hardness of the two varieties".

From determinations of specific resistance on similarly treated specimens they conclude that: "the resistivity at 0°C of tough and brittle material was identical within the experimental error".

Rogers [14] published a large number of results which showed that slowly cooled steels in the brittle condition were softer than the same steels in the quenched tough condition,

but he made no attempt in his experiments to differentiate between the effect of "extra-tempering" in the slowly cooled specimens and the effect of the change to the brittle condition. A steel which is cooled slowly from the tempering temperature is actually tempered for a longer period than a steel quenched after retaining at the tempering temperature for the same time, and would naturally be expected to be softer. Greaves and Jones, as already stated, produced evidence to show that when tempering effects are equalised there is no softening in the brittle state.

Thermal curves, taken by the differential method, have been published by Rogers (14) and by Greaves and Jones (9). No very definite conclusion could be drawn from these curves but they appeared to point to the possibility of an absorption of heat on heating and an evolution on cooling through the "brittleness range".

Guillet (15) has more recently carried out work on dilatation and thermo-electric power. His conclusions are as follows:-

"Dilatation experiments did not furnish any information," and "Thermo-electric tests gave equally no indication".

The magnetic properties of tough and brittle steel appear to have been compared in only one case, by Kayser.(16) He gave figures showing that the remanence of one steel, which was susceptible to brittleness, increased 40 per cent on slow cooling.

Figures given by Gebert (17) of magnetic tests on a nickel and on a nickel-chromium steel, quenched and tempered at various temperatures, followed by air-cooling, showed that the residual

induction in both steels increased very considerably after tempering at 425°C and above. He believed that this change was connected with temper-brittleness but he did not compare specimens in the tough and in the brittle condition.

Similar experiments by Dowdell [18] on a number of carbon and special steels showed a similar rise in residual induction to that observed by Gebert, but only in a few steels was the rise at all pronounced, and the change was much greater in all his carbon steels (0·7 to 1·2 per cent carbon) than in any of the special steels examined. He did not compare brittle and tough steels.

It is evident from the work of Dowdell that the rise in residual induction after tempering to 400° C or 500°C is very marked in steels which are not susceptible to temper brittleness. Results published by other workers [19,20,21] support this view. It seems that this particular magnetic change cannot be taken as a sign of susceptibility to temper-brittleness, but it does not necessarily follow that it is entirely unconnected with that phenomenon.

It is thus clear that all the evidence published up to the present time points to the absence of any substantial physical change accompanying the transition from the tough to the brittle condition.

EXPERIMENTAL METHOD for SPECIFIC VOLUME and BRINELL HARDNESS DETERMINATIONS.

The main experimental method used in the first part of this investigation was that of heat treatment of suitable specimens *in vacuo* followed by determinations of the specific volume and the Brinell hardness.

Specimens about 25 grammes in weight were heat-treated in a wire-wound electric furnace, the tube of which was of transparent fused silica. A high vacuum was maintained by means of a Toepler mercury pump of large capacity. It was found possible to heat-treat polished specimens at high temperatures for long periods without producing more than a slight tinting of the surfaces of the specimens, due to oxidation. Later it is shown experimentally that no alteration took place in the composition of the specimens even after many long heat-treatments.

The specimens were thoroughly ground all over and finished on 1 G French paper before the determination of the specific volume. The Brinell marks were completely removed between each treatment, but it may be said that Brinell marks produced no appreciable difference in the specific volume in a number of determinations which were made on specimens before and after Brinelling.

The specific volume was determined by comparison with a standard specimen of known specific volume by weighing in paraffin. By means of taking scrupulous care with every detail

of the process of determination, the method has been developed into one of great sensitivity and accuracy.

A first-class balance was used, giving very accurate weighings to the fourth place of decimals (1/10,000 gm.) The weights used were carefully standardised before the research was commenced and this was repeated from time to time throughout.

A standard specimen had first of all to be prepared and its absolute specific gravity determined. For this purpose a piece of annealed 0·48% carbon steel, weighing about 15 grammes, was taken. Its true specific gravity was determined as follows:-

(1) The specific gravity of the oil used was determined by means of a 100 c.c. specific gravity bottle at 15°C, the temperature being accurately measured by a Beckmann thermometer.

(2) The specific gravity of the standard specimen was then determined in the oil at exactly the same temperature of 15°C measured by the Beckmann thermometer.

The true specific gravity of the standard specimen having been determined the specific gravity of the experimental specimens was determined in the following manner.

The specimens, after preparation as already described, were placed in a desiccator for at least half-an-hour. Each specimen was then weighed lying on the balance pan. The zero of the balance was checked before and after each reading and any necessary correction made. The specimens were then immersed in paraffin (which was first filtered to remove any suspended matter) and left for over half-an-hour to allow the temperature to stabilise. A beaker of the same paraffin was used in which to weigh the specimens, these being suspended in the liquid by means of a cage made of platinum wire. The cage by itself was

first weighed. The standard specimen was then placed in the cage and weighed. Each specimen was weighed in the same manner, the standard specimen being weighed before and after each experimental specimen. If the standard varied in weight by more than 0·0002 gramme the readings were discarded and the weighings repeated. This trouble, due to variation of temperature of the oil, could usually be avoided by taking care in moving the beaker to avoid the heat of the hand altering the temperature of the oil, by taking precautions to avoid draughts altering the temperature of the room and by carrying out the weighings as quickly as possible consistent with absolute accuracy. The zero of the balance was checked throughout the weighings in paraffin, but it did not vary as in the case of the weighings in air.

By means of the precautions indicated it was found possible to reduce the experimental error to a very low figure. In many cases the specific volume determinations were repeated by re-weighing, particularly when what appeared to be unusual or particularly important results had been obtained and frequently the repeat determination was obtained identical or almost identical to the original in the sixth decimal place of the specific volume. In other cases the difference was greater but was never in any case (except in the case of an obvious error, which was occasionally obtained due to small bubbles having remained undetected on the foot of the specimen or on the cage) more than $\pm$ 0·00001 in the specific volume. This is

consequently given as the maximum experimental error.

Variations, due to heterogeneity of a steel obviously cannot be eliminated by this or any other method but these were largely obviated in this research by the use of the same specimens throughout a series of experiments.

From the data which was obtained as detailed above the specific volume was calculated as follows:-

Let weight of specimen in air = W_1

weight of specimen in oil = W_2

weight of standard in air = W_3

weight of standard in oil = W_4

specific gravity of oil = D_1

true specific gravity of standard = D_2

apparent specific gravity of standard = D_3

Then, specific gravity of specimen $= \dfrac{W_1 \times D_1}{W_1 - W_2}$

where $D_1 = \dfrac{D_2}{D_3}$

$= \dfrac{D_2 \times (W_3 - W_4)}{W_3}$

Thus, specific gravity of specimen $= \dfrac{W_1 \times D_2 \times (W_3 - W_4)}{(W_1 - W_2) \times W_3}$

and, specific volume $= \dfrac{1}{\text{specific gravity}}$

Seven figure logarithms were used in the calculations.

The Brinell hardness numbers were determined by an oil-pressure Brinell machine. A load of 1500 Kilogrammes was generally used as the more usual load of 3000 Kilogrammes made too large an impression and necessitated too great a loss in volume on grinding off the Brinell marks. In cases where the 1500 load did not produce a sufficiently large impression for accurate readings the larger load was used. A new Brinell ball was obtained at the commencement of the work. This was exactly 10 m.m. in diameter and remained completely spherical and unaltered in diameter throughout the entire research.

A platinum platinum-iridium couple was used throughout, standardized against steam and the melting points of pure tin, lead, zinc, aluminium, sodium chloride and copper. The standardization was checked several times but no alteration of the couple was found to have taken place. The couple was used in conjunction with a Siemens Temperature Indicator.

FIRST SERIES OF EXPERIMENTS

ANALYSIS OF STEELS

Steel	C. %	Mn. %	Si. %	P. %	S. %	Ni. %	Cr. %	Description
A 22	0·48	0·18	0·11	low	low	Nil	Nil	Plain carbon
N P 1	0·31	0·75	0·112	0·018	0·029	3·64	Nil	Nickel, low phosphorus
N P 5	0·33	0·70	0·065	0·098	low	3·66	Nil	Nickel, high phosphorus
A W 22	0·33	0·52	0·177	0·016	0·014	3·48	0·76	Nickel-chromium
A W 23	0·37	0·54	0·149	0·013	0·012	4·55	1·14	Nickel-chromium

N P 1 and N P 5 are two of the steels used by Andrew and Green [11] in their work on the effect of phosphorus on the notched-bar impact value. N P 1 has a very low susceptibility to temper-brittleness, whereas N P 5 has a high susceptibility. A W 22 and A W 23 were available in very small quantity only, and impact tests could not be carried out on these steels. They were both assumed to be susceptible in greater or lesser degree to temper-brittleness. The plain carbon steel was included for purposes of comparison.

Two sets of specific volume specimens of the above five steels were treated as follows:-

First Set. Oil-quenched from 850°C and tempered for three hours at various temperatures from 450°C upwards and quenched in water.

Second Set. The same treatment, except that tempering was followed by slow cooling at a rate of 2°C per minute instead of

water-quenching. This rate of cooling was adopted in order to have this preliminary series of experiments completed within a reasonable time. It was thought to be sufficiently slow to produce some change in the steels, although a slower rate of cooling would probably be necessary to produce the maximum degree of brittleness.

Quenching in oil from 850°C preceded each tempering treatment.

The specific volume and Brinell hardness results after each treatment are shown in Table I and in ~~Graphs~~ Figs. 1,2,3,4, and 5.

Steel A W 23 (~~Graph~~ Fig. 5) which contains the most nickel and chromium, shows a very pronounced and consistent change from 550°C upwards in both hardness and specific volume, but in the other steels no notable difference is apparent between the quenched and the slowly cooled specimens, except possibly in the other nickel-chromium steel, A W 22 (~~Graph~~ Fig. 4), which shows indications of a change in specific volume only, similar to the larger change observed in Steel A W 23. No change is evident between the quenched and slowly cooled specimens of Steel NP 5, which has been shown to be highly susceptible to temper-brittleness. The microstructures of the specimens of Steel A W 23 were examined after the treatments in the brittleness range in which the large change in hardness and specific volume was observed. Etched in 1% nitric acid in alcohol the slow-cooled specimen appeared to have a finer general structure than the quenched specimen and this difference was made more apparent after

electrolytic etching in 5% hydrochloric acid. The type of structure and the apparent distribution of the carbide were, however, identical. Attempts to etch the specimens in boiling sodium picrate, in order to show up the distribution of the carbide, were unsuccessful.

It is evident from the curves obtained for the carbon and the nickel steels (~~Graphs~~ Figs. 1, 2 and 3) that quenching from the tempering temperature up to about 680°C has no appreciable effect on the hardness or on the specific volume. The possibility that a strain effect induced by quenching is alone responsible for the very marked changes in Steel AW23 is completely discountenanced by the entire absence of any such change in the carbon steel or in the nickel steels, as well as by the magnitude of the change in Steel AW23 itself and by the manner in which the changes become apparent suddenly between 500°C and 550°C.

In order to find whether the extra heating received by the slowly cooled specimens had produced any relative difference between them and the quenched specimens, the ten specimens were resubjected, at the end of the series of experiments, to the original tempering treatment of three hours at 450°C. The figures for all the treatments at 450°C are given in Table 2.

The two specimens of Steel AW23 were also resubjected to a six-hours treatment at 670°C, both specimens being quenched from the tempering temperature. These results are also shown in Table 2. They clearly prove that no appreciable alteration in composition had taken place in any of the specimens, and

incidentally show the excellence of the method of heat treatment in an efficient vacuum.

The extra three hours' tempering at 670°C given to the specimens of Steel AW23 shows that the effect of extra-tempering on slow cooling must be very small. Three hours' extra-tempering at the high temperature has given only a drop of 8 Brinell and 0·000022 in the specific volume.

It is also shown in the Second Series of Experiments, described below, that extra tempering could not be made to explain the pronounced difference obtained on slow-cooling Steel AW23. The conclusion that this difference in specific volume and in hardness is due to a radical physical change which takes place during cooling was thus well substantiated.

This contraction and softening of the steel in slow cooling is what would be expected to result from the separation of a solute which had been in solution at the tempering temperature and which is retained in the latter condition on quenching the steel in water. This will be more fully discussed later after further experiments have been described.

At the time when these experimental results were first obtained it was thought that Steel AW23 must possess an unusually high susceptibility to brittleness or rather that the physical change which produces temper-brittleness must be particularly large in this steel, and that in Steel AW22 the change producing brittleness was so small as to be difficult of detection. The apparent absence of a physical change in Steel NP5 which has a high susceptibility to brittleness was

not thought to be surprising, since it has always been considered possible that brittleness induced by phosphorus may be of a different nature to that ordinarily found in low phosphorus special steels.

One or two other points in connection with the results of this First Series of Experiments call for mention at this stage. It will be noticed that the specific volume of the nickel steels (~~Graphs~~ Figs. 2 and 3) falls very considerably as the tempering temperature is raised and that this fall increases rapidly at the higher tempering temperatures. In comparison the hardness changes at the higher tempering temperatures are insignificant. Benedicks (22) has given figures to show that troostite and pearlite have the same specific volume, but the results at present referred to, together with many other results obtained in the course of this research, make it appear certain that the state of division of the carbide has a considerable effect on the specific volume. Micro-examination has shown that in the nickel steels globularisation of the carbide is exceeding rapid as compared with carbon steels, and the effect of aggregation of the carbide particles into globules leads evidently to the closest packing arrangement of the mixture of carbide and iron. Any solubility of the carbide in α-iron would certainly increase with temperature, and would oppose the observed fall in specific volume, so that the latter seems to be due entirely to alteration of the state of aggregation of the carbide. The rapid change in rate of globularisation with slight variation in tempering

temperature above 600°C is probably responsible for the variable impact figures obtained in nickel steels which has been noted by Griffiths.(12) In the nickel-chromium steels (Graphs 4 and 5) an exactly similar effect to that observed in the nickel steels is apparent but it is not nearly so pronounced. This difference can be attributed to the difference in composition and in nature of the carbides present in the two types of steels. In steels containing nickel only, that element appears to be present entirely in solid solution, the carbide remaining as pure iron carbide, but when chromium is present it combines with some of the carbon, with the result that double carbides of iron and chromium exist in the steel.

It will be seen from ~~Graphs~~ Figs. 2, 3, 4, and 5 that the effect of globularisation is so great in the nickel and in the nickel-chromium steels at high temperatures that even where slight solution of carbide in γ-iron at 680°C or 690°C leads to an increase of hardness on quenching, the specific volume still falls.

This series of experiments was continued to higher temperatures in connection with work which is described in Part II of this Thesis. The results in the Ac_1 critical range will not be specially dealt with at present. It may be pointed out, however, that the results for Steel AW23 in the critical range show that, even when considerable solution in γ-iron is induced at the tempering temperature, slow cooling through the brittleness range still gives a large fall in specific volume and in hardness, similar to that obtained at lower temperatures.

SECOND SERIES OF EXPERIMENTS

In order to be able to differentiate clearly between changes in specific volume and in hardness due to the effects of tempering alone and those due to the transition from the tough to the brittle condition this series of experiments was carried out. Greaves and Jones, as has been previously noted, equalised the tempering effects in a number of their experiments by using double heat treatments, i.e. they first treated two specimens, which they wished to compare, to obtain one in the tough condition and the other in the brittle; they then retreated the specimens, the tough being made brittle and the brittle one tough. This is a perfectly sound method but has the great disadvantage of making the heat treatment doubly laborious. If the double heat treatment were adopted in every case when one was dealing with the present methods of obtaining specific volume determinations it would result in making an already slow process intolerably protracted, when dealing with a large number of experiments. Also when various different rates of cooling are used in making physical comparisons, as in a number of experiments to be described later, it is practically impossible to apply the method of double heat treatment. Apart from these experimental considerations it was desirable to obtain data in regard to the effect of extra-tempering of nickel chromium steels.

The following steels were taken:-

Analysis

Steel	C. %	Mn. %	Si. %	P. %	S. %	Ni. %	Cr. %	
AH	0·31	0·47	0·310	0·015	0·021	4·46	1·41	
NC	0·31	0·57	0·145	0·026	0·029	3·20	0·83	
NR3	0·36	0·34	0·275	0·016	0·021	1·90	1·15	
AW22	0·33	0·52	0·177	0·016	0·014	3·48	0·76	) As in) First
AW23	0·37	0·54	0·149	0·013	0·012	4·55	1·14	) Series

The three new nickel-chromium steels, AH, NC, and NR3 were obtained, through the kindness of Dr. McCance, of the Clyde Alloy Steel Co., Motherwell, in sufficient quantity to allow of impact testing to correlate with the physical determinations.

Specimens of these steels were tempered (after oil-quenching from 850°C) at 650°C for two hours and quenched in water. They were then retempered at 650°C for additional periods of six hours each and quenched in water after each treatment. The specific volume and the hardness were determined after each treatment, and the results are given in Table 3, and ~~Graph~~ Fig. 6.

The fall in specific volume and in hardness is small in all the steels for short periods of extra tempering. Six hours of extra-tempering at 650°C produce only a difference of approximately 20 Brinell and 0·00004 in the specific volume.

From the results of tempering experiments already discussed (see ~~Graphs~~ Fig. 4 and 5) it is certain that the effect of extra-tempering is much greater at 650°C than at lower temperatures during slow-cooling, and thus it is apparent that any pronounced contraction or softening which is observed in slowly-cooled nickel-chromium steels cannot be attributed to extra-tempering. The curve for Steel AW23 shows conclusively that the change observed on slow cooling in the First Series of Experiments (~~Graph~~ Fig. 5) is not due to the effect of extra tempering but to a physical change which takes place during cooling.

This series of experiments was carried much further than originally intended, owing to an apparent very slight rise in specific volume after some of the specimens had been subjected to the first few treatments. After this rise was first noticed each succeeding treatment merely seemed to accentuate it, except in Steel N C, which shows a rise in one treatment only. The hardness dropped slightly but continuously in all the steels until the last treatment, when a very slight rise was observed in each case. The hardness results show that the rise in specific volume is not due to graphitisation of the carbide which would have produced much greater effects. It will be noticed that Steel A H shows the rise in specific volume more clearly and consistently than the other steels. It will be shown later that this same steel shows another similar change in a more pronounced degree than the other four steels. The evidence is at present probably too meagre to justify any definite explanation of this observed effect but it is consistent

with the idea that solubility of carbide in α-iron, or rather in ferrite containing nickel and other elements in solid solution, can be increased by prolonged soaking at high tempering temperatures.

THIRD SERIES OF EXPERIMENTS

As the nickel-chromium steels A H, N C, and N R 3 were now available in sufficient quantity for impact tests, experiments were carried out to find the susceptibility of these steels to temper-brittleness. At the same time the opportunity was taken to find the effect on the impact properties of tempering at temperatures above those generally used. Treatments at temperatures close to the Ac_1 range have naturally been avoided by previous workers, but in the present investigation into the physical properties it was advisable to have impact results for comparison with physical determinations made after heat-treatment at these high tempering temperatures.

Bars of the steels A H, N C, and N R 3 were first oil-quenched from 900°C and then tempered for two hours at 650°C. The same oil-quenching and tempering treatments were carried out on other bars of the same steels at 670°C, 680°C, and 690°C. The treatments were carried out in a large electric tube furnace which had an effective heating length of about 2 feet. The tube was wound "differentially", i.e. the wire was spaced closer at the ends than in the centre in order to give as even a temperature as possible from end to end. The five-inch bars

which were placed in the centre of the furnace, were thus very evenly heated throughout their length, and tests showed that the variation in temperature did not exceed a few degrees. Bars of each steel were quenched from each tempering temperature, and other bars were cooled at a very slow rate (0·3°C. per minute) in order to produce the maximum brittleness. Standard Izod test-pieces were then machined from the bars, impact tests carried out, and the Brinell hardness taken at each fracture end on the impact pieces.

The mean results are given in Table 4, and plotted in ~~Graphs~~ Figs. 7,8 and 9.

It will be seen that each of the steels has a high susceptibility to temper-brittleness. The most interesting feature of the results is the rise in impact value of the bars water-quenched above 650°C in Steels N C and N R 3 (Steel A H falls at 670°C., but might be expected to rise in the same way between 650°C and 670°C) This rise in impact value is accompanied by a fall in hardness due to the increased tempering effect as the temperature is raised. These experiments show that the higher the tempering temperature (provided the start of the Ac_1 critical range is not reached), the higher the impact value of the tough steel. Whenever the start of the allotropic change is reached, however, the impact figure falls again, and this fall is accompanied by a rise in hardness, showing that some solution of carbide in γ-iron had taken place at the tempering temperature, with subsequent formation of some martensite on quenching.

FOURTH SERIES OF EXPERIMENTS

Now that a definite physical change had been shown to take place in one nickel-chromium steel (Steel AW23) togetner with indications of a small similar change in another (Steel AW22), the next stage was to attempt to produce the maximum physical change in these steels and in others. For this purpose ten steels of widely varying composition were taken, as follows:-

Five nickel-chromium, A H, N C, N R 3, A W 22, A W23

One plain carbon A 22

Two nickel, low and high phosphorus, N P1, and N P5.

The full analysis of the above steels has already been given.

One chromium, A A X (C,0·32; Mn,0·91; Si, 0·145; P,0·026; 5,0·027; Cr, 0·97 per cent)

One high carbon, nickel-chromium, G 2 (C,1·50; Mn,0·26; Ni, 3·46; Cr, 1·80 per cent)

Specimens of these steels were treated with a view tc producing the maximum amount of brittleness in order to obtain the maximum physical change. The results of various experiments are shown in Table 5, in which figures for all the steels (except G 2) in the quenched tough condition are also given for comparison (Treatment 1) . Steel G 2 was quenched in oil from 1100°C and the others from 850°C before each treatment.

The steels were first treated in the generally accepted

manner for the production of maximum brittleness, being tempered for two hours at 650°C and slowly cooled at 0·3°C per minute. The results (Treatment 2, Table 5) all of which were thoroughly checked, were very disappointing as no great change appeared to have taken place in any of the steels with the exception of a decided fall in hardness in Steel AW23 and a fall in specific volume in Steel N P 5. The specific volume of Steel AW23 was very little altered and did not show a change in any way approaching the large change previously observed in this steel in the First Series of Experiments (the change in hardness was also not nearly so large).

The specimens from Treatment 2 were retained and fresh specimens were subjected to a tempering of two and a half hours at 670°C and slowly cooled at 0·3°C per minute. The increase of toughness after tempering at this temperature found in the Third Series of Experiments pointed to the possibility that increase in the amount of the tough condition at this temperature would lead to a larger physical change on slow cooling. This experiment (Treatment 3, Table 5) did not, however, produce results any more satisfactory than had been obtained in the treatment at 650°C. With one or two minor exceptions none of the steels even showed signs of the contraction and softening which was looked for.

In order to increase, if possible, the condition of brittleness in the specimens from Treatment 3, they were re-treated at 550°C for three hours and slowly cooled at 0·3°C. per minute. The results (Treatment 4, Table 5) again showed no

pronounced change in any of the steels. The results of these three treatments compared with the results of Treatment 1, Table 5 would inevitably lead to the conclusion that no physical change - observable at least in hardness or in specific volume - had taken place in slow cooling. The absence of a change in Steel AW23 was particularly puzzling as the previous change observed in this steel appeared to have been very thoroughly substantiated. Consideration of the results after the above failures to produce a physical change on very slow cooling led to the entirely novel conclusion that an intermediate rate of cooling would show up a change whereas very slow rates would not. The rate of cooling in the First Series of Experiments had been 2·0°C. per minute which is six times as quick as the rate used in the experiments just described.

Another experiment was consequently carried out with fresh specimens of the ten steels of the present series, tempering at 670°C. for three hours to produce the maximum amount of the tough condition at the high temperature, and cooling at 2·0°C. per minute. The results are shown in Table 5 (Treatment 5). This experiment was very successful, a pronounced change being produced in all the nickel-chromium steels. Three specimens of Steel AW23 were used in this experiment and the results of the three agree remarkably. Two of these three specimens were the identical ones used in the First Series of Experiments, so that the importance of the rate of cooling in revealing the physical change was particularly well corroborated. The

hardness of these specimens of Steel AW23 does not show a change larger than in those of Treatments 2,3, and 4. This must be due to slight solution of carbide having taken place in γ-iron at the soaking temperature, resulting in the production of a little martensite on slow-cooling. Later it will be shown that in Steel AH which is of similar composition to Steel AW23 a very low point is obtained in a cooling curve after heating at 670°C. This results, of course, in a relative hardening of the steel. From the results in this part of the Thesis and from others which will be discussed in Part II it seems very clear that to produce the maximum change on slow cooling it would be necessary to treat each steel individually by first finding the exact temperature at which the Ac_1 range commences and then tempering the oil-hardened steel just below that temperature, before slow-cooling was commenced.

It should be pointed out that if any solution of carbide in γ-iron had taken place at the tempering temperature in any of the steels it would tend to produce an expansion and hardening in the steel on slow-cooling regardless of whether the Ar_1 point on cooling was low enough to produce martensite or whether it was at a temperature high enough to allow of deposition of carbide out of solution. There is consequently no question of effects which had taken place in the Ac_1 range in any of these steels having produced a contraction or softening in the moderately cooled specimens since any such effects would produce an opposing expansion and hardening.

It is not proposed at present to draw any conclusions from the results of the other steels in this series of experiments, since less is known of the effects of extra tempering in these steels, but it is interesting to note the very decided fall in hardness in the high carbon, nickel-chromium steel, C 2, after Treatment 5, which produced such a large change in the low carbon, nickel-chromium steels.

It is not surprising that the alterations in specific volume with varying rates of cooling had not been observed in previous work, since this method had hardly ever been applied, but it seemed remarkable that the changes in hardness had not been previously noticed. A search of the work published up to the present, however, showed that no hardness determinations had been made in experiments in which the effect of varying rates of cooling on the impact figure had been carried out. Later it was found that the changes in hardness are somewhat masked, in many cases, by the effects of extra-tempering when the usual tempering temperatures of 650°C or lower are used.

In view of the well established fact that the amount of brittleness is more or less directly proportional to the rate of cooling, it has previously been assumed that the physical change which produces the change in impact value would also be proportional to the rate of cooling. It is now evident from the present work that two changes giving opposite effects, depending on the rate of cooling, may take place in slowly cooled steels.

Also the fact that no appreciable change is revealed in density or in hardness when the tough state is compared with the fully brittle has led, naturally enough, to the false assumption that no change would be observable when intermediate rates of cooling were used.

The present work makes it apparent that the similarity in density and in hardness of the tough and brittle states is merely accidental, and that in fact a considerable transformation takes place in the physical condition of the steels on slow cooling.

It is now clear that many of the results which have been previously attributed to the softening effect of extra-tempering must have been due, in part at least, to the change taking place with intermediate rates of cooling. For example, Rogers,[14] using a rate of cooling of 1°C per minute, obtained a softening in all his steels, in some cases the softening being considerably more than could be attributed to extra-tempering. The experience of other workers at that time led to doubt being openly cast on the accuracy of his results in the discussion which followed his paper. Greaves and Jones,[9] using a rate of cooling of 0·3°C. per minute and equalising tempering effects, found no difference in hardness between the water-quenched and the fully brittle condition. These results, which have previously been regarded as incompatible with each other, are now shown to be due to the difference in the rate of cooling employed, as the rate used by Rogers will be, for many steels at least, an intermediate rate. The experimental results of

these investigators agree completely with the observations made in the present investigation.

FIFTH SERIES OF EXPERIMENTS

As a result of the discovery that the rate of cooling has an important bearing on the production of changes in specific volume and in hardness, the present series of experiments was carried out in order to find the effects produced by various cooling rates.

The following seven steels were taken:-

Three nickel-chromium, A H, N C, and N R 3

One plain carbon, A22.

Two nickel, low and high phosphorus, N 3 (C,0·29; Mn,0·65; Si, 0·113; P,0·036; S,0·032; Ni,3·22 per cent) and N P 5.

One chromium A A X.

The analysis of the steels, with the exception of N 3, has been given in previous series.

Several of the steels which were used in the first part of this research were originally available in very small quantity, only a few inches of bar being available in some cases. Steel AW23 had unfortunately to be omitted from this series, together with AW22 and N P 1, as sufficient of the steel was not available for all the experiments. N P 1, was replaced by a fairly low phosphorus nickel steel, N 3.

Five sets of these seven steels were treated, after a preliminary oil quenching from 850°C, as follows:-

First Set - Tempered for two hours at 650°C and quenched in water

Second " - " " " " " " " slowly cooled at 3°C per minute

Third " - " " " " " " " " " " 2°C per minute

Fourth " - " " " " " " " " " " 1°C per minute

Fifth " - " " " " " " " " " " 0·6°C per minute

All the steels except N 3 had been treated in the Fourth Series of Experiments at a cooling rate of 0·3°C per minute (Table 5, Treatment 2). The results for all the various rates of cooling are given in Table 6, and plotted in ~~Graphs~~ Figs. 10 to 16 inclusive.

The carbon steel provides a very useful comparison for the others. Rate of cooling has no effect whatsoever on its specific volume, while the hardness falls slightly and practically continuously due to extra-tempering.

The changes in the special steels are seen to be comparatively small and somewhat erratic but that they are none the less real is seen by the consistent absence of any changes in the carbon steel. It is apparent from the results of the nickel-chromium steels that the previous conclusion that the changes after tempering above 650°C are accentuated is fully borne out. Some of the results of the treatments at 670°C in the Fourth Series of Experiments are included in ~~Graphs~~ Figs. 11, 12, and 13 to illustrate this point.

The effects of extra tempering have somewhat confused the changes due to brittleness. As the cooling rate becomes slower the amount of extra-tempering becomes greater especially in the two or three slowest treatments, in which the amount of

extra-heating at temperatures above 600°C becomes great. This results in a contraction and softening which opposes the expansion and hardening which evidently takes place in the production of the fully brittle state. When this is kept in mind it is seen that the previous conclusions in regard to the effect of varying cooling rate are well confirmed.

It can be easily seen from the small changes in hardness how these have remained undetected by previous workers, since the tempering temperatures usually employed were either 650°C or lower.

The most interesting and significant result of this series of experiments is the revelation in the nickel steels and in the chromium steel of changes similar to those found in the nickel-chromium steels. The high phosphorus nickel steel, N P 5, has been shown (11) to have a high susceptibility to temper-brittleness, and it shows changes both in specific volume and in hardness (~~Graph~~ Fig 15). Steel N 3 (~~Graph~~ Fig 14) shows a slight change in hardness only. Its susceptibility is not known, but is probably low. The chromium steel is known to be susceptible to brittleness, and it shows decided changes in hardness (~~Graph~~ Fig. 16). Extra-tempering may have masked a probable accompanying change in specific volume.

That changes of exactly the same physical nature are involved in all these steels of widely varying composition is an inevitable conclusion from these results, and that these are the physical changes involved in temper-brittleness is scarcely open to doubt. It is now clear that a considerable

transformation must take place in the physical condition of steels which are susceptible to brittleness when they are slowly cooled from the tempering temperature.

It may also be said here that the condition of high hardness and high specific volume in water-quenched steels, as compared with the condition of low hardness and low specific volume in steels cooled at a moderate rate, is not evidently essential to reasonable toughness in the steels. The moderate rates of cooling which produce a considerable contraction and softening in the steels may produce only a slight fall in the impact value. It is probable then that the toughness of a steel is not dependent nearly so much on any special condition of the steel which, in itself, gives toughness, as on the absence of the brittle condition which is induced by very slow cooling rates. The toughness is, of course, dependent, as has been previously shown, to a relatively small extent on a special condition giving added toughness to a steel by water-quenching.

SIXTH SERIES OF EXPERIMENTS

The five nickel-chromium steels A H, N C, N R 3, A W 22 and A W 23, were used in this series of experiments to find the effect of tempering temperature on the specific volume and hardness.

Specimens of the steels were first oil-quenched from 850°C. and then tempered for six hours at 300°C. and quenched in water. The specific volume and the hardness were determined and the tempering repeated (without intervening

oil-quenching) at 400°C. The experiments were continued at various temperatures up to 670°C., six hours soaking being given at each temperature, except at 670°C., where three hours soaking was given in each of two successive treatments. The specimens were water-quenched after each treatment at the tempering temperature, but no intervening oil-quenching was carried out; the specimens were thus continuously tempered at a series of progressively rising temperatures.

The results are given in Table 7, and Figs. 17 and 18.

With the exception of steel A H no outstanding change is shown in any of the curves. It had been thought that a slight rise in specific volume might be observed as the temperature was raised through the brittleness range. No rise has, however, taken place, but it is evident from some of the curves, notably steel A H, that, whereas the hardness falls continuously, there are "halts" on the specific volume curves. The hardness falls continuously and evenly in all the steels, but the specific volume does not. From 450°C to 650°C in Steel A H the specific volume is almost unaltered in spite of a softening of 100 Brinell. A slight similar halt is seen in the other steels at about 600°C to 650°C.

It has been previously noticed that Steel A H showed a more marked expansion on prolonged tempering at 650°C (Fig.6, Second Series of Experiments) than was shown by the other nickel-chromium steels. Similarly in the present series this steel shows a more pronounced halt than the others. It is thus probable that this steel develops the physical change

which is connected with temper-brittleness to a greater degree than the others. Increase in the amount of the tough state as the temperature is raised through the brittleness range would account for the effect observed in Steel A H, where the specific volume, while it does not rise, opposes the contraction which one would expect from the extra-tempering received by the steel.

The steels, in general, show a larger fall in specific volume between 650°C and 670°C than between any of the immediately preceding temperature intervals. This effect must be due to globularisation at high temperatures and is further proof that the state of division exerts a decided effect on the specific volume of a steel.

The curves also give further proof of the comparatively small effects of extra-tempering in nickel-chromium steels at high tempering temperatures. It will be noticed that the extra three hours' heating at 670°C gave a very small fall in all the steels.

SEVENTH SERIES OF EXPERIMENTS

The detection of the physical changes in specific volume and in hardness which has now been fully described did not immediately render the cause of temper-brittleness obvious. No similar changes had ever been previously observed in metallurgy or in any branch of physical science, as far as the present author was aware. Consequently it was not possible to state the cause since there was no scientific data which

would explain the observed effects. The author felt, however, that the changes could be explained in a certain manner which will be discussed later, and the present series of experiments was carried out with a view to establishing an analogy to the effects observed in the brittleness range.

In connection with some experiments which will be fully described in Part II of this Thesis, a result was obtained in a nickel steel which, it was thought, might throw light on the matter.

The nickel steel concerned was Steel N 3, which has been used in the Fifth Series of Experiments. Bars of this steel had been heated at 695°C until complete solution of carbide in γ-iron had taken place. Certain bars were quenched in water after this treatment and others were slowly cooled at 0·3°C. per minute to 400°C. Tensile and Izod test-pieces were machined from the bars and mechanical tests carried out. Examination of a micro-specimen, which had been placed in the furnace along with the bars to follow the heat treatment, showed that a large proportion of the carbide of the slowly cooled steel had separated out in a remarkably complete film round the grain boundaries. A micrograph of this is shown in Fig 20 (considerable difficulty was found in obtaining anything like a good photograph of this structure owing to the glare of light from the surface of the polished specimen) It was noticed also that the hardness of the micro-specimen was unusually high - considerably higher than would be expected in an annealed steel of this composition.

When the Izod impact test-piece, which had been subjected to the same treatment as the micro-specimen, was broken, it was found that the impact value varied in a manner not obtained in any of the many other bars subjected to other treatments. When the hardness at the fracture faces was taken it was found to vary also from one end of the bar to the other, and to be higher at all points than what was expected in the annealed state.

It was thought that the effect of variation in physical properties of this steel might provide the desired analogy to the behaviour of special steels in the brittleness range. The test-bar and the micro-specimen were consequently subjected to a number of experiments.

A diagrammatic sketch of the Izod impact test-piece referred to is shown in Fig 19. The manner in which the impact value and the hardness varied from one end of the bar to the other (and inversely to each other) is clearly seen.

The following experiments were carried out:-

(a) Microsections were prepared at each of the fracture faces, and the polished sections were boiled in an alkaline solution of sodium picrate to etch up the carbide. It was found that the amount of carbide round the grains varied along the length of the bar in a similar manner to the impact figure and the hardness. As the impact value decreased and the hardness increased the boundary became more apparent. The original micro-specimen, which had a higher hardness than any part of the test-piece, had a more complete boundary film than any part of the latter. This variation of the amount and

completeness of the boundary in the different specimens was clearly beyond doubt. This observation was confirmed by an independent observer, who placed the specimens in the same order as the author.

The grain-size of each specimen was estimated and found not to vary. The possible influence of variation in grain size was thus eliminated.

(b) Specific volume test-pieces were then taken from each fracture end of the test-bar and from the original micro-specimen, and their specific volume determined. The results are given in Table 8, Experiment (b). It will be seen that the specific volume was also much higher than would be expected in the annealed state. Except in Specimen 2 it varies with the hardness.

(c) The four specimens were then tempered at 650°C for fifteen minutes and cooled quickly in the furnace. The specific volume and the hardness after this treatment (Table 8 Experiment (c)) were both greatly lowered, and micro-examination showed that the grain-boundary was largely broken up into small globules. This was the only apparent change in structure and the difference in properties seemed to be attributable only to the difference in state of distribution of the carbide.

(d) The specimens were re-tempered at 650°C.for one hour, and this treatment resulted in a further decrease in specific volume and in hardness (Table 8, Experiment (d)) and the grain-boundary was completely destroyed. The arrangement of the larger globules which had been formed by globularisation from

the carbide of the boundary indicated faintly in parts the original granular structure.

(e) Three of the specimens were then quickly annealed by heating to 760°C. for a few minutes and cooling quickly in the furnace. The results (Table 8, Experiment (e)) show that the specific volume and the hardness are much lower than when the carbide was obtained at the boundaries in the same specimens — Experiment (b). It would be expected that the steel cooled much more quickly and from a higher temperature above the Ar_1 point would show a higher hardness and possibly a higher specific volume. The opposite, however, is obtained in the case of Experiment (b) and (e) Table 8. The steel cooled extremely slowly has much the higher specific volume and hardness, and the only apparent physical difference between the two structures produced is in the distribution of the carbide round the grain-boundaries in the slowly-cooled steel and the production of the usual mixture of ferrite and pearlite in the other.

A tensile test-bar in which the carbide was distributed as a network round the grains gave a tensile strength of 56·4 tons per square inch and an elongation of 11·0 per cent. The figures supplied by the makers for this steel were 55 tons per square inch tensile strength and 20-25 per cent elongation (oil-hardened 820°C. and water-tempered 620°C). The existence of the carbide grain-boundary has produced little or no difference in the tensile strength. The elongation is lowered by half, but the difference in Izod values is much greater. The

average Izod value for the grain-boundary bar was 9·8 foot-pounds, whereas in the oil-hardened and water-tempered condition it was given as 60 foot-pounds. There is thus the strong indication that the distribution of the carbide at the grain boundary has a much greater effect on the impact value than on the other mechanical properties.

EIGHTH SERIES OF EXPERIMENTS

While it was felt that the analogy which has been dealt with in the Seventh Series of Experiments had been apparently well established, there yet remained a possibility that the high specific volume and high hardness in the grain boundary nickel steel might be due to other causes which were not apparent. The manner in which the grain boundary of this steel had been formed was also not entirely clear and the present Series of Experiments was undertaken in order to attempt to clear up these points.

The heat treatment to which the grain boundary nickel steel had been subjected, namely, one week's heating in the Ac_1 range at 695°C. followed by very slow cooling, was an unusual one. It was thought that more normal treatments, at temperatures clearly above the Ac_1 range, followed by slow cooling, might be made to produce the grain boundary. Also it was felt that if this same effect could be produced in a plain carbon steel it would strengthen the analogy.

The experiments detailed on Table 9 were consequently carried out. The specimens treated, with the various heat treatments, results, and microscopical observations are indicated fully on the Table. Full discussion of these experiments need not be undertaken here since the results are almost entirely negative, as far as the production of a grain boundary is concerned. The treatments used did not produce a carbide grain boundary in any of the steels in any way approaching that obtained previously in the nickel steel.

These experiments, however, together with others described in Part II of this Thesis, served to make quite clear the manner in which the carbide grain boundary had been formed in the nickel steel Izod bar. It had been thought that the slow cooling used after tempering had been the principal factor in producing the boundary by expelling the carbide as the grains were being formed or re-formed on passing through the Ar_1 range. It now became clear that this was erroneous, and that the chief factor was the state induced in the steel at the high temperature before cooling was commenced.

What had taken place in the nickel steel was clearly as follows:- the treatment had been carried out at the temperature at which, in this steel, the formation of γ-iron entectoid solution on heating is almost complete, i.e. practically all the carbide had been brought into solution in γ-iron and the remainder of the steel still consisted of α-iron. The long treatment had caused the grains of α-iron to grow to

some extent (but the final grain-size was still quite small, see Fig.19), and the γ-iron entectoid solid solution containing all or nearly all of the carbide had become expelled or segregated round the grains thus really forming the matrix of the steel at the soaking temperature. When cooling took place at the end of the soaking treatment the carbide separated out from the boundary of γ-iron solid solution in the Ar_1 range. The α-iron formed in the pearlite would tend to join the main grains of α-iron, thus keeping the carbide at the boundaries.

It is shown in Part II of this Thesis that Ar_1 must have taken place at practically the same temperature as would be obtained in an ordinary annealing treatment from say 850°C. Thus the possibility that a lowering of the normal Ar_1 had taken place as a result of the unusual treatment was eliminated. Ar_1 having taken place at the normal temperature the slow rate of cooling would in the ordinary way have produced a softening in the steel relative to a normalising or annealing treatment. This would oppose the expansion and hardening effect produced by the grain boundary. When this is considered the results obtained are all the more significant.

The uncertainty of production of a grain boundary is shown in the way in which the latter varied in amount along the length of the Izod test piece. Other experiments will be described later in which the effect of heating within the Ac_1 range was to globularise the areas of γ-iron entectoid solid solution formed at the treatment temperature. When this takes place the α-iron remains the matrix and as no γ-iron solid

solution is at the boundaries of the α-iron grains no carbide boundary would be formed on slow cooling. The difference in these two effects of soaking in the Ac_1 range is not quite clear and is in any case not a material matter at present, but the undoubted difference which does exist accounts for the variation in the amount of carbide boundary along the length of the Izod test-bar.

An interesting point in connection with the properties of carbon steels was revealed by this series of experiments. It will be seen from Experiments 2 and 3, Table 9 that whereas the relative hardness of the 0·22% carbon steel and the 0·48% carbon steel is as would be expected after annealing and tempering treatments, yet the values for specific volume are the reverse of what would be expected. This is a peculiar anomaly which is not at present understood. It may be connected with the tendency shown by low carbon steels to expel the carbide, to some extent to the grain boundaries but, if so, one would expect the hardness to show a change similar to the specific volume. Thermal curves of these two steels showed that they were quite normal (Figs. 21 and 22.)

NINTH SERIES OF EXPERIMENTS

Some experiments were carried out on the quenched Izod bar of the nickel steel, N 3. This bar had been subjected to the same soaking temperature of one week at 695°C but had been quenched from the tempering temperature. Specimens were cut from each fracture end in the same way as in the grain boundary bar, Fig.19, and, together with the quenched micro-specimen, these were subjected to the heat treatments shown on Table 10 in which the results of specific volume and hardness determinations are given. It will be seen (Experiment 1) that while the hardness of this bar is considerably greater than in the grain boundary specimen the specific volume is much lower. This is due to the fact that the trapping of a good deal of the entectoid γ-iron solid solution as austenite has kept the specific volume low but that evidently sufficient martensite was formed throughout the steel to keep the hardness high.

This is the only case in which the specific volume of a steel quenched from a temperature in the Ac_1 range has been found to be less than the specimens slowly cooled from the same temperature. It will be seen from Figs. 1,2,3,4, and 5 that a large number of such treatments have been carried out on carbon, nickel, and nickel chromium steels and in none of these cases was the specific volume of the slow cooled specimen higher than that of the quenched.

This gives consequently further confirmation of the expansion effect produced by the grain boundary in Steel N 3.

The results of the other experiments on Table 10 show

practically nothing which throws light on the points presently at issue. They will not, therefore, be further discussed here.

SPECIFIC RESISTANCE

TENTH SERIES OF EXPERIMENTS.

The importance of rate of cooling in showing up changes in specific volume and hardness having been discovered it was advisable to make determinations of other physical properties in a similar manner.

Bars of the three nickel-chromium steels A H, N C, and N R 3, which were available in quantity, and which all showed a high susceptibility to temper-brittleness, were used for resistivity work.

Bars of over 25 cms. in length were necessary for magnetic determinations (described later) and bars of the necessary length of the above steels were included in the following heat-treatments which were given to the resistivity pieces.

As the effect of extra-tempering on the resistance and magnetic properties of steels was quite unknown it was thought advisable to equalise, as far as possible, the tempering effects in the heat treatments. When different cooling rates are used this is impossible without using a large number of bars, but a very good approximation to it can be made when the method here used is employed.

The bars were heated in an electric tube furnace which is described in Part II. Pieces of bar 9 cms. long were used for the resistivity work. The required number of pieces were treated as follows:-

(1) All bars were oil-hardened from 850°C.

(2) Pieces of each steel were:-

(a) Tempered 2 hours at 660°C and slow cooled at 0·3°C per minute.
(b) Re-tempered 2 hours at 660°C and quenched in water.

This gave the steels finally in the tough condition.

(3) Pieces of each steel were:-

(a) Tempered 2 hours at 660°C and quenched in water.
(b) Re-tempered 2 hours at 660°C and slowly cooled at 2·0°C. per minute.

This gave the steels finally in the intermediate condition.

(4) Pieces of each steel were:-

(a) Tempered 2 hours at 660°C and quenched in water.
(b) Re-tempered 2 hours at 660°C and slowly cooled at 0·3°C. per minute.

This gave the steels finally in the fully brittle condition.

The number of necessary heat treatments was reduced by combining Treatments 3a and 4a and Treatments 2a and 4b.

The pieces of bar were machined down to 3 mm. diameter for the determination of their resistivity.

EXPERIMENTAL METHOD

The method adopted for determination of the electrical resistances was of the same type as that used in determinations of specific volume, namely, by comparison with a standard specimen of known resistance. This type of experimental method is much superior to methods for the direct determination of the absolute physical characteristics of each steel. Comparative results can be obtained to a degree of accuracy which is not approachable by the direct method.

One of the resistance pieces was selected to act as a standard. The absolute resistance was determined as follows:-

The test-piece was held firmly by powerful elastic bands on a set of two knife edges, which were screwed to a wooden base at a known distance apart (about 5 cms) and exactly parallel with one another. A current of about 1·5 amperes from a battery of Ni-Fe cells was passed through the specimen and the difference of potential across the specimen at the knife-edges was measured on a Tinsley Vernier Potentiometer. A shunt of known resistance was placed in the circuit, and the current was accurately determined before and after each determination on the specimen by measuring the difference of potential across the shunt. A thermometer placed with its bulb in contact with the specimen gave the temperature of the latter. During a number of days it was found possible to take a number of readings at temperatures which varied a degree or two on each side of 20°C. At each temperature sets of readings were taken from the specimen, the latter being moved a

millimetre or two along the knife-edges between the sets of readings. From the average of each set of readings the resistance between the knife edges was calculated from the current and the difference of potential between the knife edges. The diameter of the specimen was accurately determined by a micrometer, readings being taken at cross diameters each half centimetre along the specimen and averaged.

From these data the specific resistance at each temperature of the standard specimen was calculated and from a graph the specific resistance at 20°C was read off.

The resistivity of the other specimens was then determined by comparison with the standard. The unknown specimen was placed on another set of knife-edges in series with the standard. Readings were then taken of the difference of potential across the unknown, readings across the standard being taken before and after. This was repeated for as many points as necessary on the unknown specimen. The diameter of the unknown was determined in the same way as described for the standard. The specific resistance was then obtained from the following calculation.

$$\text{Specific resistance, } \rho = \frac{E_1}{E_2} \times \frac{d^2}{4} \times R \times \frac{\pi}{e}$$

where E_1 = D.P. across unknown

E_2 = D.P. across standard

R = absolute resistivity of standard

d = mean diameter of unknown

e = distance between knife-edges

In this method it is necessary to assume that the temperature coefficient of resistance is the same for the different steels and for the different conditions. A number of other experiments proved that this factor could make no appreciable difference in the accuracy of the results. In all the experiments the temperature of the specimen was very near to the arbitrary standard temperature of 20°C.

The mean results are shown in Table 11. It will be seen that a small but quite definite fall has taken place in the resistance in the slowly cooled specimens. The slower the cooling rate the larger is the fall of resistance, but the difference between the tough and the intermediate state is greater in each case than is the difference between the intermediate and the fully brittle state. These experimental observations do not agree with the results of Greaves and Jones, (9) who concluded that there was no difference in resistivity between the tough and brittle states. This conclusion was, however, based on very little data, as only two of their published readings really compare the brittle and tough states, and, also, the tempering temperature which they used was the low one of 600°C. It has been clearly shown in the first part of this paper that the physical change which accompanies brittleness is considerably accentuated after tempering at higher temperatures than this.

It will be seen that the total fall in resistance between the tough and the fully brittle states (Table 11) varies in the different steels. This total fall is proportional to the nickel content of the steels.

ELEVENTH SERIES OF EXPERIMENTS.

In the Third Series of Experiments the three nickel-chromium steels A H, N C, and N R 3 were treated at various high tempering temperatures to give the tough and fully brittle conditions. From the longer ends of the Izod test-pieces from these experiments resistivity specimens were machined and determinations carried out on them. The pieces were fairly short - less than 5 cms - and a distance between knife edges of 3 cms. was the maximum which could be used. The specimens were thus turned down to 2·5 mm. diameter and this gave results of the same order as in the last series of experiments. The resistivity was determined in exactly the same manner as already described and the results are given in Table 12.

These results fully confirm the results obtained in the last series. The fall in resistivity from tough to brittle condition is greater. This must be due to the effects of extra-tempering, which were not equalised in these experiments. It will be seen that as the tempering temperature is raised the percentage drop in resistivity between the quenched and the slowly cooled specimens of Steel A H rises continuously. but that in Steels N C and N R 3 it first falls and then rises. If reference is made to Figs. 7, 8, and 9 it will be seen that a rise in the percentage drop is associated with a fall in impact value of the quenched specimens, and a fall in percentage drop with a rise in impact value. It is clear that the percentage drop falls until the start of the Ac_1 range is

reached and that then it rises due to the production of martensite in the quenched specimens. It might appear from the fall in percentage drop with rising temperature before solution in the Ac_1 range commences that the amount of tough state decreases with temperature. This is the opposite of previous conclusions from the specific volume and hardness work. The effect on the resistivity must be explained by the assumption that increase in globularisation with rising temperature lowers the resistivity in greater proportion than the increase in tough state raises it. From the previous observations on the marked globularisation at these temperatures it is felt that this is very likely to be the correct explanation.

MAGNETIC PROPERTIES

EXPERIMENTAL METHOD

The magnetic properties were determined by the bar and yoke ballistic method. The general lay-out of the apparatus is shown in Fig. 23. The specimen, which is in the form of a bar 25 cms. long by 1 cm. in diameter, is clamped firmly in a yoke of soft annealed wrought iron. Primary and secondary coils are wound round the specimen in the yoke. The magnetising current is supplied by a sufficient number of accumulators and its strength is regulated by an adjustable resistance, R_1, and measured by the drop of potential across a shunt of known resistance which is placed in the circuit.

A Cambridge and Paul Testing Set was used for this purpose. The reversing Key, K, has two of its terminals connected through an adjustable resistance, R_2, which may be shorted by a switch S, as shown. When S is closed throwing over the reversing Key K reverses the current without altering its strength. When S is open the reversing key not only reverses the current but changes its strength by an amount depending on R_2. For permeability tests this Key S is kept closed but it is required for taking cyclic B-H curves. The reversing Key, C, allows the current to be sent either into the primary coil on the specimen or into the primary of the mutual inductance. The latter serves as a means of standardising the ballistic galvanometer. When a known current is reversed in the primary of the mutual inductance and the ballistic effect of the reversal observed the number of lines of induction for each scale division is calculated as follows:-

$$B \text{ (for one division)} = \frac{I \times 10^6}{A \times T \times d.}$$

where I = current in amperes

A = cross sectional area of specimen in square centimetres.
T = number of turns in secondary coil on specimen.
d = throw of ballistic galvanometer in scale divisions.

A resistance placed in the secondary circuit allows the sensibility of the ballistic galvanometer to be regulated so that a suitable maximum throw is given in the experiments.

The short circuiting key across the galvanometer is kept closed except when it is desired to allow the swing of the galvanometer to be read. The two secondary coils of the specimen and the mutual inductance are permanently connected in series, so that the secondary circuit undergoes no change when the magnetising current is sent into one for testing or the other for standardising.

The ballistic galvanometer, lamp, and scale were rigidly fixed in position before commencing the determinations. Thus any possible error due to the determination of B for one division was eliminated.

Tests were made for the maximum induction and residual induction at values of H from 0 up to as high a value of H at which a steady current was available from the batteries.

Cyclic B-H curves or hyseresis loops were then taken for a maximum value of H equal to 150. This value was arbitrarily fixed as being high enough to give almost the maximum value of B rem in all the specimens and at the same time using as high a current (1·72 amperes) as could be kept running steadily from the batteries for reasonably long periods.

The B max and B rem determinations were made as follows:

The current was adjusted to give a desired value of H and the Key, K, reversed several times to establish a cyclic state. The switch, S, was kept closed throughout. The short circuiting switch across the galvanometer was then opened and the ballistic throw on reversing the Key, K, observed.

This gave the throw for B max. B max = $\frac{1}{2}$ T_1 x B for one division. Where T_1 = observed throw. The Key was then reversed several times and thrown over towards the top of the diagram, Fig. 23; the resistance, R_2, was then disconnected, the switch, S, suddenly opened, and the ballistic throw observed. This gives the throw for a step down from the known value of H to H = o and from it B rem is obtained by difference.

B rem = $(\frac{1}{2} T_1 - T_2)$ x B for one division, where T_2 = observed throw.

The above was repeated for all desired values of H.

H is calculated as follows:-

$$H = \frac{4\pi}{10} \times \frac{N}{l'} \times I$$

where N = number of turns on primary of specimen.

l' = clear length of specimen in yoke in centimetres.

I = current in amperes.

For determination of the hysteresis loop the following was the procedure adopted:-

The current was first adjusted to give the desired maximum value of H = 150. The Key, K, was reversed as before and thrown over towards the top of Fig. 23. The resistance, R_2, was then adjusted to the desired value, the switch, S, suddenly opened, and the ballistic throw observed. This gives the effect of a step down from H = 150 to a value of H depending on the resistance, R_2. S was then closed, a cyclic state re-established, R_2 set to a larger value and a larger step similarly made by opening S. In this way points from

H = 150 to H = C were determined.

To complete the curve it is necessary to determine points for which H is negative. This was done by using the reversing Key, K, to pass from the full positive current to a less negative current. After reversing the Key, K, several times with the switch, S, closed and the full current on for H = 150, the Key, K, was switched from top to bottom while S was held open, thereby introducing R_2 as well as reversing. This was repeated for a number of diminishing values of R_2 to the opposite extremity of the cycle. B and H were calculated from the observed values of current and throw as before. In this way the complete hysteresis loop could be drawn.

TWELFTH SERIES OF EXPERIMENTS.

Specimens of Steels A H, N C, and N R 3 had been treated, as described in the section on Specific Resistance, to produce the tough, intermediate and fully brittle conditions in each steel. Test pieces about 30 cms. in length were machined down to 1 cm. in diameter for use in the yoke which was available. The exact mean diameter of the specimens was determined in a similar manner to that used in the case of the resistivity specimens.

The magnetic properties of these specimens were determined by the method described above.

The results are given in Tables 13 to 22 inclusive and in Figs. 24 to 32 inclusive.

The tough specimen of Steel A H was sent to the National Physical Laboratory for determination of its magnetic properties so that it could be used as a standard. The figures obtained by the author agreed very closely with those given by the N.P.L. test; at the higher values of H the authors figures for B max. were larger than those given by the N.P.L. possibly due to the yoke used. Proportionate corrections have been applied by comparison with the standard and the corrected figures are given in the Tables.

The individual test figures are given in Tables 13 to 21 and in Figs. 24 to 32. Table 22 contains a summary of the more important magnetic characteristics from the hysteresis loops so that the chief results can be easily compared.

It will be seen that the property which gives the most distinct variation from tough to brittle condition is the remanence or residual induction. Kayser [16] found that in one steel examined the remanence increased 40% in the slowly cooled state. This is a very large rise and, from the present work, it would seem that a large part of it may have been due to the extra-tempering of the slowly cooled specimen. That there is a very appreciable difference in remanence due to the brittleness change alone is, however, definitely shown by the present results, where tempering effects were equalised. In Steels N C and N R 3 the total rise from tough to brittle condition is just over 10%. The author would have expected it to be greater than this in Steel A H, but it appeared from later work that at the tempering temperature of 660°C - which

was chosen so that the maximum brittleness change would be induced in the steels - slight solution of carbon in γ-iron takes place at the beginning of the Ac_1 range. This would give effectswhich would oppose the rise in remanence observed and it seems certain that a larger rise in remanence would be obtained after treatment at a slightly lower temperature in this steel.

The influence of rate of cooling on the rise in remanence is erratic in the different steels. This is only to be expected since the arbitrarily chosen cooling rates do not necessarily represent the correct rates for the production of the intermediate and the fully brittle conditions in each steel. Thus in Steel N C most of the change has evidently taken place at the rate of $2{\cdot}0^{\circ}C$ per minute, whereas only a small portion has taken place in this treatment on Steel A H and about two-thirds has taken place in Steel N R 3.

Of the other magnetic properties the only other one which has been found to vary directly from tough to brittle condition in each steel is the susceptibility. The maximum susceptibility increases as the cooling rate is made slower. The coercive force decreases from tough to brittle but Steel N R 3 is apparently an exception to this. The maximum induction does not vary greatly in the different conditions, and the area of the hysteresis loop could not be said to show a definite change.

The remanence of the steels is thus evidently the magnetic property which best shows up the change from the tough to the

brittle condition. The rise in this property has been very thoroughly confirmed by both the series of experiments of which the results are given. At values of H beyond 150 the remanence rises very slightly in each steel and attains a constant value. These maximum values of the remanence bear the same relation to each other as those taken from the cyclic curves.

THEORETICAL CONSIDERATION OF RESULTS

It has been clearly proved that in steels susceptible to brittleness a moderate rate of cooling (2°C.or 3°C. per minute) produces a considerable fall in specific volume and in hardness. The moderate rates of cooling which produce this physical change may only produce comparatively small variations in impact value.

It is evident that the magnitude of this physical change is roughly proportional to the degree of brittleness to which the steels are susceptible. Thus in nickel-chromium steels, which have a high susceptibility, the change in specific volume and in hardness is very marked, whereas in nickel and in chromium steels which have a low susceptibility the change is small. As the phosphorus content of a nickel steel is raised the change is accentuated.

This physical change may be explained by the deposition, on moderately slow cooling, of a solute which is in solid solution in the steel at the tempering temperature, and which is retained in solid solution by quick cooling. The only

possible solute which is present in all the steels in sufficient quantity to give the pronounced physical changes is carbide (either simple carbide of iron or double carbide of iron and chromium).

Impurities which are present in the steels may act along with the carbide, but in themselves could not give rise to the large changes observed, particularly in the nickel-chromium steels.

It is thus suggested that a certain amount of carbide is present in solid solution in α-iron (which may be either nickel or chromium ferrite) at high tempering temperatures, and that this carbide is, partially at least, deposited in a fine state of division throughout the mass on moderately slow-cooling.

The suggestion of Rogers [14] that brittleness is due to solution and redeposition of carbide is thus regarded as being partially true, but it is necessary to assume a further change in order to explain fully the results obtained in the present investigation and also to show the real cause of the lowering in impact value.

The amount of carbide held in solid solution at the high temperature may vary considerably with the composition of the steel, and also may increase considerably with increase in temperature, especially near to the Ac_1 range. This certainly seems to be indicated by the rise in impact value found in the nickel-chromium steels above 650°C.

The deposition of carbide from solid solution will not, however, in itself produce any great degree of brittleness in a steel, and it is evident from the present work that another physical change is induced in the steels by very slow rates of cooling. It has been clearly proved that very slow rates of cooling give a considerable increase in the specific volume and in the hardness when compared with the values obtained by moderate cooling rates. This increase in specific volume and in hardness cannot be explained merely by a further deposition of carbide when the rate of cooling is slower, although it is probable that the slower the rate of cooling the more complete will be the deposition of carbide from solid solution.

It is believed that the state of distribution of the carbide must be the predominating influence in this increase in specific volume and in hardness, and also in the high degree of brittleness which accompanies it.

The evidence provided by the experiments on a nickel steel containing grain-boundary carbide (Seventh and other Series of Experiments) points very strongly to the fact that, when the carbide is distributed round the grain boundaries, the specific volume and the hardness are considerably raised over the values obtained for these properties in the same steel with the same amount of carbide present but distributed more or less homogeneously throughout the mass. The grain-boundary condition results also in the production of brittleness in the steel, whereas the steel in its normal quenched and

water-tempered condition is tough.

These experiments give, therefore, a complete analogy to the physical changes observed in steels susceptible to temper-brittleness and provide a reasonable explanation of the variations in impact value.

It is believed then that the rise in specific volume and in hardness on cooling very slowly from the tempering temperature is due to the expulsion of the carbide, which is deposited out of solution, to the grain-boundaries. If a complete network of carbide is formed during cooling it is not difficult to see how it would affect the physical properties to such a marked extent. The actual grains of the steel will become virtually separated from each other by the network of carbide, and the co-efficient of contraction of the latter will exercise a determining influence on the final volume of the mass. From work published by Honda (23) it is clear that the co-efficient of expansion of the carbide in plain carbon steels is less than that of the iron. This is almost certainly true of any carbide in special steels. Consequently when a boundary network is produced a greater volume of the mass will be obtained on slow cooling due to the lower co-efficient of contraction of the carbide. The effect of the carbide network on the hardness is clearly explainable by its interference with slip or flow of the mass of the steel under pressure. No such interference takes place when the carbide is evenly distributed throughout the mass.

Microscopical examination of the steels used in the present work has not yet produced complete confirmation of the above theory. It is probable that any grain-boundary which is produced is extremely small in magnitude and therefore difficult of detection. The grain-boundary which was found in the nickel steel (Fig. 20) was easily detected by etching in sodium picrate, but was extremely difficult to pick out by ordinary etching - in fact, its existence could not have been established at all had it not been possible to etch up the carbide in the polished specimen. The grain-boundary in this case must be much larger than any which is produced in the tempering zone in steels susceptible to temper-brittleness. In the nickel-chromium steels, which have high susceptibilities, it has not yet been found possible to etch up the carbide in the polished specimens.

In the high-phosphorus nickel steel, however, the polished specimen was etched by sodium picrate and an unmistakable grain-boundary was revealed. This was exactly similar in structure, but not so apparent, as the grain-boundary shown in Fig. 20. A faint indication of a similar boundary was obtained in the low-phosphorus nickel steel.

The effect of the state of aggregation of the carbide (apart altogether from the effect of its distribution round the grain-boundaries) may be briefly commented on.

The results have shown that the fall in specific volume in carbon steels with rise in tempering temperature is very small, but that it is very pronounced in nickel steels and also

appreciable in nickel-chromium steels. This effect must be due to globularisation of the carbide, which micro-examination has shown to be slow in carbon steels but which becomes very rapid in nickel steels as the temperature is raised near to the Ac_1 critical range.

It has been pointed out by other investigators (24,25) that globularisation of carbide can only be explained by assuming a certain solubility of the carbide in α-iron. The rate of globularisation will then depend on the degree of solubility of the carbide in α-iron (other factors, such as initial state of division and temperature, being equal). It is thus probable that the increased rate of globularisation in a nickel steel is a proof, in itself, that the addition of nickel increases the amount of carbide which is held in solid solution in the ferrite at higher temperatures. As the rate of globularisation increases until the Ac_1 critical range is reached, so also must the amount of carbide present in solid solution increase. Similarly, manganese may increase solubility in the same way as also other elements in solid solution, such as chromium in excess of the amount necessary to combine with the carbon of the steel.

It has been shown in the experiments on specific resistance and magnetic properties that these properties do not show the double effect which is revealed in density and in hardness. The author had expected that the double effect might be apparent in the resistivity but that was not apparent from the results of the experiments carried out. The fall in

resistivity was found to be greater the slower the rate of cooling. It is possible, however, that the production of a network of carbide throughout the mass may exercise an opposing effect to the effect which is obtained from the separation of carbide out of solid solution. At lower temperatures the double effect might account for the apparent similarity of resistance in the tough and brittle states found by Greaves and Jones.(9)

The magnetic properties would be expected to show variation in direct proportion to the amount of carbide which separates out of solution, the grain boundary having no effect. It has been shown that the magnetic properties vary directly with the rate of cooling.

Quantitative interpretation of the results is not easy but it is quite possible that the fact that the fall in resistance was found to be directly proportional to the nickel content gives a clue of importance. It has been previously shown that many of the results can only apparently be explained by increase in solubility of carbide in ferrite with the addition of nickel, particularly at high tempering temperatures. Kase (26) estimates that carbon is soluble in pure nickel up to 0·25% at room temperature and that the solubility increases to 0·55% as the temperature is raised up to the entectoid of the carbon-nickel system (1318°C.) It is thus likely that the presence of nickel in solid solution in α-iron increases the solubility of carbide. Other elements in solid solution in α-iron may act in the same manner, and it seems very possible that the resulting increased

solubility at high temperatures and re-deposition on slow-cooling by the manner described above is the true explanation of temper-brittleness.

SUMMARY

Physical determinations have been made on a wide range of steels, including carbon, nickel, chromium and nickel-chromium, with a view to detecting physical changes accompanying the well-known change in impact value found in many special steels and known as temper-brittleness.

As a result of the work the following physical changes have been proved to take place:-

(1) A double change in specific volume and in hardness. At moderate rates of cooling (2°C. or 3°C. per minute) from the tempering temperature steels susceptible to brittleness show a contraction and a softening. At very slow cooling rates (0·6°C. or 0·3°C. per minute) an expansion and a hardening takes place relative to the condition found after moderate cooling, the specific volume and the hardness returning to values very similar to those obtained in specimens quenched from the tempering temperature.

(2) The contraction and the softening obtained on moderately slow cooling is accentuated as the tempering temperature is raised, provided the start of the Ac_1 range is not reached.

(3) The specific resistance of slowly cooled steels is

lower than that of the same steels quenched from the tempering temperature. No double effect is observed as in the case of the specific volume and the hardness, the decrease in resistivity being greater the slower the rate of cooling.

(4) Of the magnetic properties the remanence or residual induction shows the most pronounced change. It rises as the rate of cooling becomes slower. Again there is no sign of a double effect.

A theory has been formulated to account for the phenomenon of temper-brittleness. This is based on the above physical observations and is supported by experimental evidence on a steel in which the carbide had been obtained segregated largely at the grain-boundaries. The theory may be briefly stated as follows:-

The introduction into a steel of nickel, or certain other elements which form solid solutions with α-iron, increases the solubility of carbides in ferrite, especially at higher tempering temperatures. On quenching from the tempering temperature the carbide dissolved in the ferrite is retained in solid solution. On cooling at a moderate rate the carbide in solid solution at the high temperature separates out, at least partially, in a fine state of division throughout the mass. On cooling at a very slow rate the carbide which separates is expelled to the grain boundaries thereby producing a brittle network of carbide throughout the mass.

A number of other points in the behaviour of carbon and special steels have been dealt with.

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PART II

Ac_1 and Ar_1 in Special Steels.

INTRODUCTION

It has been noted by Brayshaw,[1] Brearley,[2] Jones,[3] and Greaves [4] that partial hardening of certain special steels is brought about by tempering at temperatures considerably below the normal Ac_1 maximum (as obtained on a thermal curve) followed by quenching in water. This effect has been frequently recognised in practice and is commonly guarded against in the heat-treatment of some special steels by the use of relatively low tempering temperatures. Carpenter, Hadfield, and Longmuir (5) drew attention to the fact that in thermal heating curves of nickel steels the Ac_1 change is revealed by an absorption of heat over a range of temperature commencing well below the Ac_1 maximum. Jones,(6) from thermal experiments on this subject concluded that Ac_1 in alloy steels takes place over a range of temperature and he explained this as being due to the progressive solution of a mixture of isomorphous carbides present in alloy steels. This general explanation seemed to the author to be decidedly weak, since the great mass of evidence points to the existence of one simple carbide in many special steels where the Ac_1 range is very pronounced, as for example in certain nickel steels. In other special steels, known to contain more than one carbide, as, for example, chromium steels where the ratio of chromium to carbon is less than 4·3:1, there was a possibility that Jones' explanation might hold.

But while the work of Jones showed experimentally that in many special steels the Ac_1 change was not completed at constant temperature, no attempt has been made to define the conditions which govern the extent of the Ac_1 range or to elucidate the reactions which take place in it. It was the aim of the present work to throw light on this matter. Most of the experimental work has been carried out on the Ac_1 range but it was felt that success in this direction would necessarily provide information which would simplify the interpretation of other and apparently more complex changes in special steels, as, for example, the double cooling points in nickel and in nickel-chromium steels.

SPECIFIC VOLUME

EXPERIMENTAL METHOD.

It was thought that some information might be gained on the reactions in the Ac_1 range by the application of a new method to the problem. A method involving the determination of specific volumes was consequently adopted in the first experiments.

Suitable specimens of the steels, about 25 grammes in weight, were heat-treated in vacuo in a wire-wound electric furnace. In the earlier experiments a thick vitreous silica tube was used and selected tubes of this material kept a high vacuum excellently below about 700°C but above that temperature their efficiency in this respect was found to fall away with rising temperature. Later experiments were carried out in tubes of transparent fused silica and these were found perfect at all temperatures. Polished specimens could be treated at high temperatures in these tubes for long periods with only a slight tinting of the surfaces by oxidation.

The merit of this method in completely avoiding any alteration in composition during treatment has already been shown in Part I of this Thesis.

A platinum platinum-iridium thermocouple was used throughout and this was frequently checked against the melting point of pure aluminium (658°C.) which was near to the temperatures used in the most important heat treatments. The e.m.f. recorded by the couple did not alter in the slightest degree

in spite of many hundreds of hours of heating in vacuo.

The temperatures were read on a Cambridge and Paul Universal Test Set in the earlier experiments and on a Siemens Temperature Indicator in the later. In the soaking experiments in the critical range the retention of the temperature to within a variation of a few degrees over six consecutive hours was found very trying and the null method with galvanometer spot thrown on a large scale would have been preferred for this type of work. The latter method was not, however, available for this work at the time when it was carried out.

The temperature was controlled by means of a sliding resistance, to within the slightest possible variation of the millivoltimeter pointer, taken as a maximum as equal to $\pm$ 2°C. to 3°C. The mean soaking temperature was taken as 2°C. to 3°C. below that desired so that the temperatures shown in the Tables and Figs. are the maximums reached in each case.

The specimens were prepared and the specific volume determined in the manner fully described in Part I of this Thesis.

FIRST SERIES OF EXPERIMENTS

Three nickel steels with constant nickel and varying carbon were chosen for this series, together with a plain carbon steel for purposes of comparison.

Analysis of Steels

Steel.	C.	Mn.	P.	S.	Si.	Ni.
A4	0·71	0·16	low	low	0·10	nil
2	0·26	0·27	0·014	0·025	0·103	3·64
4	0·44	0·33	0·015	0·038	0·125	3·65
6	0·87	0·41	0·013	0·025	0·149	3·64

Specimens of each of these steels, about 25 grammes in weight, were first quenched in oil after half-an-hour at 850°C. They were then tempered as follows:-

Treatment	1	2	3	4	5	6	7	8	9	10.
Temperature, °C	540	540	590	640	660	675	685	695	700	705
Period of Heating in Hours.	12	an extra 12	12	12	12	12	12	12	12	12
								11	12	13
								705	710	715
								an extra 12	12	12

The lengthy treatments were given so that complete stability would be produced at each temperature, and thus eliminate from the results any doubt that might arise on this

point. Each twelve-hour treatment was made up of two periods of six hours each and rapid quenching in cold water followed each treatment. There was no intervening treatment between the various experiments, and in heating up for each treatment the temperature was raised as quickly as possible without overshooting the desired soaking temperature. Any appreciable redistribution of carbide or other constituents due to the tempering out at lower temperatures of solid solutions formed in the previous treatment was thus effectively prevented.

The specific volume and Brinell hardness results after each treatment are shown in Table 1 and in Figs. 1,2,3 and 4. The microstructure was examined in all cases and photomicrographs taken where thought desirable. Some of the latter are shown in Micros. 1 to 17 inclusive.

The experiments were discontinued after Treatment No.13 as the specimens had become rather small for accurate work owing to the necessity of grinding off Brinell marks after each experiment.

The carbon steel, Fig. 1, shows a softening as the temperature is raised but little or no change in specific volume. The microstructure of the carbon steel showed progressive globularisation as the temperature was raised.

Figs. 2, 3, and 4 show that when solution in γ-iron first takes place in the nickel steels at 685°C an expansion and a hardening is obtained in the quenched specimens and this increases with time and temperature. But when the temperature is still further raised a contraction and a softening sets in

and this is followed by a second expansion and hardening at still higher temperatures. The first expansion and hardening is associated with a marked duplex structure which is shown in Micros 1 and 2 and also in Micro 18, which is from a separate experiment. The white constituent in the microsections was seen to increase with carbon content and also as the soaking temperature was raised. It is thus evident that it is this constituent which causes the expansion and hardening. These white areas existed at the soaking temperature as γ-iron solid solutions, but quenching has failed to retain them as austenite and it is concluded that they consist of an unstable solid solution of carbide or carbon in α-iron containing nickel. It will be shown later that the concentrations of these areas vary but that they are in general relatively high in nickel and low in carbon when only a part of the carbide has been brought into solution. It was clearly seen in all the specimens that the carbide dissolved progressively as the temperature was raised. Micro 1 shows the white areas formed at the soaking temperature together with undissolved carbide which was blackened by boiling in an alkaline solution of sodium picrate. As the temperature of soaking is raised the well-defined duplex structure breaks down and Micros. 3,4,8,9,11 and 12 show indefinite broken-up structures which are the result of a re-arrangement of equilibrium.

The fall in specific volume and hardness which follows the first rise in the curves is associated with the formation of another well-defined duplex structure which gradually

separates out from the broken-up conglomerate obtained in intermediate stages. This structure is shown in Micros. 5, 10, and 14 in the three different nickel steels, and its gradual formation can be followed by comparing the micrographs of each steel after previous treatments. It is evident from the curves that austenite is being trapped on quenching and as the globularised constituent in Micros. 5,10, and 14 clearly varies with the carbon content it is concluded that this constituent consists of austenite and the ground mass of ferrite.

As the temperature is further raised this constituent disappears into the ground mass and less austenite and more martensite are formed due to the gradual dilution of the γ-iron areas at the soaking temperatures.

Special attention may here be drawn to Micro. 3 which throws light on the manner in which the carbide grain boundary structure in the nickel steel, N 3, referred to in Part I of this Thesis, was obtained. The γ-iron areas have, as shown in Micro. 3, segregated to the grain boundaries after this treatment,thus producing a condition which will give a carbide grain boundary on slow cooling. It will be seen from the other Micros. that this low carbon nickel steel has a tendency to form this structure, whereas the higher carbon nickel steels have not. Micro. 15 shows the structure of Micro. 3 at a lower magnification and similarly Micros. 16 and 17 show the structures of Micros. 8 and 11 respectively. Micro. 15 reveals a pronounced granular structure which is not

apparent in the other two cases.

The unusual nature of the results obtained in this series rendered it advisable that they should be confirmed by other experiments and a further series of similar treatments was carried out on fresh specimens of the same steels. It was thought that better quantitative results would be obtained by shorter and unbroken treatments at regular intervals and this was found, as shown below, to be the case.

SECOND SERIES OF EXPERIMENTS

Fresh specimens of the steels used in the First Series were first oil-quenched from 850°C.

They were then tempered at 670°C. for 6 hours and quenched in water. This tempering was repeated at intervals of 5°C. for periods of 6 hours each up to 705°C. after which the temperature intervals were the same but the time of each soaking was reduced to 3 hours. The steels were quenched in water after each treatment. Hardness determinations were not made as it was clear from the First Series of Experiments and from many other results obtained that after quenching treatments in the Ac_1 range the hardness followed the specific volume in all cases except when graphitisation occurred, as in one case in this investigation, namely in Steel No.6. The microstructure was examined after each treatment and the previous observations were fully confirmed.

The results are shown in Table 2 and Fig. 5. The specific

volume passes through changes exactly similar to those obtained in the First Series, but the regular spacing of the successive treatments has led, as expected, to better quantitative results. There is little alteration in the carbon steel throughout the whole range of treatments. This gives an excellent contrast to the effects shown by the nickel steels. In these the expansion which commences at 685°C. and continues to 695°C. or 700°C. is directly proportional to the carbon content as is shown in Fig. 6 where the maximums in Fig.5 are plotted against percentage of carbon. Above 695°C. or 700°C. the specific volume falls in Steels Nos.2 and 4 and in these two steels the fall is again proportional to the carbon content. But the contraction which one would expect in Steel No.6 is obscured by graphitisation. In both series of experiments a small amount of graphitisation took place in this steel after all the carbide had been brought into solution, i.e. in the dip of the specific volume curve. The formation of even a small amount of graphite has a very pronounced effect on the specific volume. The graphitisation was plainly seen, on examining the polished specimens under the microscope, as lines of small dots which appeared to be situated along the boundaries of original dendrites in the steel.

THIRD SERIES OF EXPERIMENTS

It was thought that experiments like those carried out in the first two series might be applied with advantage to other steels with a view to throwing further light on the peculiar reactions previously observed. Very few special

steels of suitable composition were available at the time, but it was decided to use three nickel steels of varying phosphorus content and two nickel-chromium steels. The results from these steels would show whether phosphorus had any effect on the changes and also if the same cycle of changes was observable in nickel-chromium steels.

ANALYSIS OF STEELS

Steel	C.	Mn.	P.	Si.	S.	Ni.	Cr.
NP1	0·31	0·75	0·018	0·112	0·029	3·64	nil
NP3	0·30	0·71	0·058	0·103	low	3·68	nil
NP5	0·33	0·70	0·098	0·065	low	3·66	nil
AW22	0·33	0·52	0·016	0·177	0·014	3·48	0·76
AW23	0·37	0·54	0·013	0·149	0·012	4·55	1·14

Specimens of each of these steels were quenched in oil from 850°C. They were then tempered at 665°C. for 6 hours and quenched in water. This treatment was repeated - without intervening oil-quenching - every 5°C. up to 725°C. The specific volumes and Brinell hardnesses were determined after each treatment, and the results are shown in Table 3 and Figs. 7 to 11 inclusive.

It will be seen that as far as the effect of phosphorus is concerned the results are negative. No particular difference could be said to show itself in the behaviour of the three nickel steels. A similar cycle of changes is seen to take place as in previous experiments. The changes

are not very pronounced due to the relatively low carbon content.

In the nickel-chromium steels, Figs. 10 and 11 one might expect a greater hardening effect at the start of the Ac_1 range than in the nickel steels. The reverse, however, is obtained. Later it will be shown that in a steel of the composition of AW23, solution in γ-iron must start at about the temperature of the first treatment, 665°C. Thus the absence of hardening at temperatures just above this must be due to the fact that the γ-iron portions formed at the soaking temperature are either wholly or partially trapped as austenite, thus keeping down the specific volume and the hardness. At about 690°C or 695°C in both the nickel-chromium steels an expansion starts and continues to 715°C. At 720°C. a lowering in both specific volume and hardness of both the steels was observed. This is probably due to the same causes as in the nickel steels but it is apparent that the same pronounced cycle of changes does not take place in nickel-chromium steels, at least of the compositions used.

Other experiments in the Ac_1 range of a slightly different nature have been given in Part I of this Thesis. In Figs. 2 and 3 of Part I it will be seen that the nickel steels pass through similar changes to those already described. The difference between these experiments and those described in the present paper is that the specimens

were quenched in oil from 850°C between each tempering treatment. It is clear that the cycle of changes is similar in both types of treatment.

FOURTH SERIES OF EXPERIMENTS

MECHANICAL TESTS

It was thought that it would be of interest to find the effect of heat-treatment in the Ac_1 range on the mechanical properties of a nickel steel.

The following steel was used:-

Analysis

Steel	C.	Mn.	Si.	P.	S.	Ni.
N3	0·29	0·65	0·113	0·036	0·032	3·22

Four six-inch bars of this steel were first oil-quenched from 850°C. and then were subjected to a very drastic treatment at 680°C. to 690°C for one week. The treatment was carried out in an electric tube furnace wound "differentially" over two feet, so that the six-inch bars in the centre were kept at a very uniform temperature throughout their length. The course of the treatment was followed by quenching out from time to time a number of small micro-specimens which were placed in the furnace along with the bars.

A week at the temperature indicated only brought about a partial solution of the carbide, the microstructure showing the duplex structure obtained in this region plus a fair amount of undissolved carbide, well globularised.

At the end of the soaking treatment two of the bars were quenched in iced brine, and the other two were very slowly cooled at 0·3°C per minute.

Tensile and Izod test pieces were machined from the treated bars, and mechanical tests carried out. The results are shown in Table 4.

Four other bars of the same steel were subjected to a similar treatment at 695°C to 700°C for four days. Two bars were quenched from the soaking temperature and two were slowly cooled at 0·3°C per minute. All the carbide appeared to have been in solution at the soaking temperature as prolonged boiling of the quenched microspecimen in sodium picrate failed to show up any carbide. The slow-cooled specimen showed that a large part of the carbide had separated out at the grain-boundaries. The effect of this structure on the physical properties of the steel has already been fully dealt with in Part I of this Thesis.

Mechanical test results from test-pieces machined from these treated bars are also shown in Table 4.

Heat treatment in the range of temperature used in these experiments has two effects, namely, globularisation of undissolved carbide and solution of carbide. This gives

a double effect on the hardness when partial solution only is obtained as in Treatments 1 and 2, Table 4. Thus while a fair amount of martensite was produced in Specimen 1 yet its hardness when compared with Specimen 5 is low. The effect of the martensite in hardening the steel in this case is evident by comparison with the steel slowly cooled from the same temperature, Specimen 2.

The mechanical properties have not been greatly impaired by either of the Treatments 1 or 2 when the very drastic nature of the treatments is considered.

In the case of Treatments 3 and 4, in which all, or nearly all, of the carbide had been brought into solid solution in γ-iron at the high temperature, the mechanical properties are greatly affected. A large amount of martensite has been produced in Specimen 3, and, as would be expected, this gives a large increase in the tensile strength and poor figures for elongation, reduction of area and Izod impact.

In the bars from Treatment 4 the carbide had separated largely at the grain-boundaries. There is no doubt that all the carbide which had been in solid solution had been completely deposited by the slow-cooling treatment (the specimens were held at 430°C for 17 hours after extremely slow cooling to that temperature). It is interesting to note how little effect the distribution of the carbide has on the tensile strength.

In the ordinary way one would have expected the complete separation of the carbide in Specimen 4 to have raised the Izod value considerably when compared with Specimen 3, but, in this case, the production of brittle grain boundaries has kept it low.

FIFTH SERIES OF EXPERIMENTS

THERMAL ANALYSIS

It was evident from the specific volume and hardness experiments that the changes in the Ac_1 range must be due to redistribution of constituents. Consequently the effects of similar heat-treatments followed by thermal cooling curves might be expected to give indications of the distributions involved. Work of this kind had only apparently been carried out in one case, by Jones.[6] In this work he found in some steels a lowering of the Ar_1 point of about 20°C, after tempering in the lower part of the Ac_1 range. As his results concerned alloy steels of widely varying composition it was not expected that large effects would be revealed by the present work. It was hoped, however, that the effects of redistribution might be sufficient to throw some light on the problem. As will be shown below pre-conceived notions on this matter proved to be quite erroneous. Very large effects due to redistribution were in many cases observed and it became evident that in Jones'

work his curves had not been taken to sufficiently low temperatures.

It has been very well established by a number of investigations [5,7,8,9,10] that in nickel steels increase of carbon content has no effect on the temperature of the Ar_1 change but that increase of nickel progressively lowers that change. Thus if nickel is redistributed in a pure nickel steel at the commencement of the Ac_1 range it would be revealed on a cooling curve by a lowering of the normal Ar_1 change, whereas, if the carbide alone was concerned, no alteration in the temperature of the Ar_1 change would be observable.

The following steels were used in the thermal work to be described:-

Analysis of Steels

Steel	C.	Mn.	Si.	P.	S.	Ni.	Cr.
N 3	0·29	0·65	0·113	0·036	0·032	3·22	nil
A H	0·31	0·47	0·310	0·015	0·021	4·46	1·41
N C	0·31	0·57	0·145	0·026	0·029	3·20	0·83
N R3	0·36	0·34	0·275	0·016	0·021	1·90	1·15
A A X	0·32	0·91	0·145	0·026	0·027	nil	0·97

The thermal specimens were treated in a high vacuum so that any effects due to oxidation would be avoided. The specimens, which were in the form of cylinders about $1\frac{1}{2}$ inches long by $\frac{3}{4}$ inch in diameter with a hole drilled to the centre, were contained in a thick wrapping of asbestos paper, and the thermocouple was jammed with asbestos wool into the drilled hole with its junction tightly against the metal. The

temperature was measured and the curves taken with a Tinsley Vernier Potentiometer.

The nickel steel, N3, after oil quenching from 850°C, was tempered at 690°C (24°C below the normal Ac_1 maximum, Fig.12) for one hour and a cooling curve taken. This is shown in Fig. 13. It is seen that an evolution of heat took place over a range of temperature from 472°C to 412°C, whereas the normal Ar_1 maximum is 577°C (Fig.12). The specimen was then re-heated to 700°C for one hour, and a cooling curve again taken. This resulted in hardly any alteration in the evolution of heat obtained in the previous treatment, but brought in a new evolution at from 595°C to 534°C this being very nearly co-incident with the start and finish of the Ar_1 evolution obtained on a normal cooling curve from high temperatures. The soakings, followed by cooling curves, were repeated at 5°C intervals up to 715°C and then at 725°C, 750°C, and 800°C, with the results shown on Figs. 13 and 14. The lower heat evolution persists until the temperature is raised to 715°C but becomes progressive-ly smaller. At 715°C the lower temperature range of evolution is practically identical to that obtained in the original treatment at 690°C. At 725°C the low range is destroyed. Meantime, throughout these treatments, the high range of heat evolution increases in magnitude until at 725°C it alone remains. It is clearly seen from the curves (Fig.13) that, after the two ranges are produced at 700°C, treatment at 705°C and 710°C tends to bring them together, but that when the lower point is almost destroyed at 715°C the lower limit of the top

range rises. After soaking at 725°C a separate Ar_3 change becomes visible and is plainly revealed after treatment at 750°C, which is well above the Ac_1 range.

A number of other experiments were carried out on this steel and the curves are shown on Figs.14 and 15. After soaking at 675°C no evolution of heat was revealed. In order to be certain that this was so a curve was taken down to 40°C (Fig.15) and this shows no discontinuity. At 685°C an evolution was obtained (Fig.14) extending from 450°C to 363°C which is decidedly lower than that obtained after the treatment at 690°C (Fig.13). On soaking the oil-quenched specimen at 680°C a very low point was produced at from 400°C to 358°C. This specimen was reheated (without intervening oil-quenching) at 685°C for 4 hours and it is seen that the point produced by the previous treatment at 680°C was slightly greater. But 4 hours at 685°C did not produce nearly such a large evolution as was previously observed after one hour at the same temperature (Fig.14). This is due, of course, to the difference in the previous treatment, the higher range of heat evolution having been produced when the steel was heated immediately after oil-quenching to the soaking temperature of 685°C.

In order to find how rapid was the redistribution of constituents a curve was taken of the oil-hardened steel by heating it as in a quick thermal curve to 700°C and switching off the current just before that temperature was reached. The steel by this treatment was in the lower part of the Ac_1 range

for as short a period as could reasonably be arranged, certainly under half-a-minute. It is seen in Fig.14 that two very small points were produced on the curve, one at 582°C, i.e. slightly above the normal Ar_1 point, and a low point at 377°C.

In the specimen previously heated to 700°C (Fig.13) two points were produced on the cooling curve. That this was purely due to its previous treatments is shown by another curve (Fig.15) taken after heating the oil-hardened steel at 700°C without previous treatments at lower temperatures. Only one point was obtained, the heat evolution extending over a range of 80°C from 500°C to 420°C. The lower limit (420°C) of this evolution is the same as in the previous treatment at 700°C (Fig.13) but it commences 40°C higher and no top point is revealed.

After this treatment at 700°C a heating curve was taken and is shown in Fig.15. This curve shows a point with its maximum at 709°C, i.e. 5°C below the normal Ac_1 maximum, due to the carbide which had given the previous low point, and a higher point which is due partly to the remainder of the carbide going into solution and partly to the Ac_3 change.

The oil-hardened steel was subjected to another one hour treatment at 700°C, but this time was cooled to above the start of the Ar_1 heat evolution previously obtained, and a heating curve taken. This is also shown in Fig.15. It shows that the top point is due to changes not brought about in the previous treatment. Absorption of heat commences at a higher

temperature and the normal Ac_1 point is very small.

An experiment was carried out to show whether the top point in these curves was due to carbide not yet dissolved or to the Ac_3 change. The steel (after oil-hardening) was heated in an ordinary heating curve until the Ac_1 heat absorption had just completed itself. The current was switched off so that this change was just complete when cooling commenced. The steel was then cooled to 595°C (above Ar_1) and a heating curve taken. This showed a large top point (Fig.14) due mainly to the α to γ change (Ac_3) and a small low point due to a little carbide not dissolved in the first treatment.

It is seen from the work on this nickel steel that the Ac_1 range starts at about 680°C, i.e. 34°C below the normal Ac_1 maximum. No steels of nickel content differing appreciably from this steel were available but the nickel-chromium steels A H and N R 3 containing 4·46% and 1·90% nickel respectively were treated to find the temperature at which the Ac_1 range in them commences. In these steels the Ac_1 maximums take place at very slightly different temperatures, 734°C in the case of Steel A H (Fig.21) and 740°C in Steel N R 3 (Fig.24) As will be seen from Figs. 22 and 25, however, the Ac_1 range commences at 660°C in Steel A H, i.e. 74°C below the normal Ac_1 maximum, giving a very low evolution of heat, and in Steel N R 3 no point was evident until after treatment at 710°C (Fig.25). The extent of the Ac_1 range is thus proportional to the nickel content in these steels.

Similar experiments to those carried out on the nickel steel were carried out on a chromium steel. In this steel (A A X) all the chromium (except possibly a minute proportion) is in combination as carbide. The normal Ac_1 maximum is at 758°C (Fig.16) and treatment at 740°C failed to reveal any solution at the soaking temperature (Fig.17). At 745°C some solution took place but the point which was observed on cooling (Fig.17) was at the normal Ar_1 temperature, and no redistribution whatever was revealed by subsequent treatments (Fig.17). It is evident that the Ac_1 range is small in this steel and indeed it may be due entirely to the manganese content, which is high. Experiments on a steel free from manganese and high in chromium would be necessary to find the effect of free chromium in solution. A very faint evolution of heat is evident at lower temperatures in some of the curves on Fig.17 and it is thought that these are due to the manganese present in this steel.

Other experiments on this steel are shown in Figs.18,19, and 20. The steel was heated just to the Ac_1 maximum in the same way as has already been described in one of the experiments on the nickel steel, N 3. The steel was then cooled to 705°C (above Ar_1) and a heating curve taken from there. This curve (Fig.20) shows a very definite point slightly above the normal and must be due, as in the case of the nickel steel, mainly to the α to γ (Ac_2) change. That this is so will be seen from the cooling curve (Fig.20) taken immediately after

the treatment just referred to. It will be seen that a distinct Ar_3 point is visible. Other curves (Figs.18 and 19) show that this point is progressively lowered with rise in initial temperature.

Cooling curves of the steels A H, N C, and N R 3, which have nickel contents of 4·46%, 3·20% and 1·90% respectively, are included (Figs. 21, 23, and 24) to illustrate a matter which will be discussed later. Steel A H shows two points on cooling (Fig.21) and Steel N R 3 one point (Fig.24) while Steel N C shows an evolution over a wide range of temperature. To be sure that this curve of Steel N C was in no way accidental a repeat of the experiment was carried out and the two curves shown (Fig. 23) are plotted from data from two separate experiments. It will be seen that the repeat reproduced the exact shape of the first curve.

It may be noted here that in a number of Dejean's curves[9] there are indications of re-distribution effects similar to those described in connection with the nickel and nickel-chromium steels in the present research. In a high speed tungsten steel containing 18% W, 0·7%C, 4% Cr and a little vanadium his cooling curves show a small heat evolution at temperatures considerably below the normal Ar_1 point, after the steels had been heated to temperatures in the Ac_1 range. With a steel containing 0·39%C, 2·44% Ni., 1·83% Cr, 1·72% Cn. he gets similar results. In the case of each steel the lowest heat evolution obtained on cooling took place after the lowest

tempering treatment at a temperature in the Ac_1 range, and as the tempering temperature was raised through the Ac_1 range the low temperature evolution occurred at progressively rising temperatures, finally disappearing.

These curves can be seen in Figs.1 and 2 of Dejean's paper.(9) They give clear evidence of redistribution in the Ac_1 range. The heat evolutions are small, in comparison with the other points which were the subject of Dejean's research, but are nevertheless quite distinct on the curves published. Dejean made no reference whatsoever to them.

THEORETICAL CONSIDERATION OF RESULTS

It is an inevitable conclusion from the experimental results that redistribution of the special element which is present in solid solution is the salient cause of the Ac_1 change taking place over a range of temperature. It has been shown in the specific volume and hardness experiments that variation of carbon content in nickel steels has no effect either on the temperature at which the Ac_1 range commences or on the apparent extent of that range. As the carbon content is increased the changes which take place are merely accentuated in proportion to the carbon content, no modification of the temperature range being produced.

From the thermal curves on nickel and nickel-chromium steels it has been clearly shown that the temperature at which the Ac_1 range commences depends on the nickel content of the

steels. The temperatures at which solution in γ-iron commences is rapidly lowered by increase of nickel. Further, it is clear that the amount of redistribution which takes place in the γ-iron first formed depends upon the amount of nickel originally present in solid solution in the ferrite.

In the chromium steel examined by the thermal method no redistribution due to the chromium (which is combined as carbide) is revealed, but the manganese present in solid solution gives indications of effects similar to those of nickel. (The amount of lowering of Ar_1 for each 1% of manganese will be roughly $2\frac{1}{2}$ times that for the same amount of nickel) The temperature range of Ac_1 in this chromium steel is evidently small and it may be entirely due to the manganese present. It is possible, however, that in steels of low chromium content there is selective solution of the carbides, as suggested by Jones [6]. In the steel used there may be two carbides, Fe_3C and $Fe_3C.\ Cr_3\ C_2$, [12] and these may be subject to selective solution when the Ac_1 change commences. If this does take place it can be explained by consideration of equilibrium in the ternary system, but for the present it is sufficient to conclude that in steels in which all the chromium is combined as carbide the temperature of Ac_1 is small.

Where, however, a special element is present in solid solution, e.g. nickel, manganese, and chromium above the 4·3% Cr: 1%C ratio, the Ac_1 change may take place over a range of temperature, provided there is a difference in solubility of the special element between α and γ-iron.

The effect of nickel in this connection will be specially dealt with here, since in almost all of the steels used it is the element particularly concerned. Similar considerations must, however, apply to other similar elements.

It is necessary to refer to the system iron-nickel-carbon in order to show clearly what takes place in the Ac_1 range. A general diagram has been drawn by Kase (13) but it is necessarily, in the present state of knowledge, incomplete and takes no account of the changes at present under consideration.

The portion of the diagram which is required for this work can, however, be constructed from previously known data, together with data from the present paper. This has been done and a diagrammatic view is given in Fig.26. This diagram is given merely as an illustration of the points presently at issue and other points have been purposely omitted from it.

Scott (10) finds that in steels up to 3·5% nickel increase of nickel lowers the Ac_1 point (maximum) by 10·5°C for each 1% of nickel added and he calculates that the carbon concentration of the entectoid becomes smaller by 0·042% for each 1% of nickel. These figures agree well with those of other workers.

Carpenter, Hadfield, and Longmuir (5) give figures for the temperature of Ac_1 maximum in steels up to 15% nickel.

From the above figures the mean line A B has been drawn in Fig.26. This line starts at the entectoid of the iron-carbon system and slopes downwards towards increase of nickel and decrease of carbon. But this line only represents the

maximum of the Ac_1 change at entectoid compositions and it is necessary in order to represent the Ac_1 range to have two lines diverging from the iron-carbon entectoid point. The lower line A C represents the start of the Ac_1 change in steels of entectoid composition and the top line A E the completion of this change. To include steels of all carbon contents two planes generated horizontally on both sides by these lines make the diagram, in this respect, complete. A vertical section through the line of entectoid compositions (A B, Fig.26) is shown in Fig.27. Some of the results of Carpenter, Hadfield and Longmuir and of Scott have been plotted as indicated. A mean line through the maximumsof their curves is shown dotted, A B. The first-named investigators used a differential method and the maximumsof their curves agree well with Scott's. The differential method gives, however, the apparent start of the Ac_1 range at temperatures which are undoubtedly much too low. The line A C, representing the start of the Ac_1 range has been drawn through Scott's figures, taken by the inverse-rate method. This line includes the point D, which has been shown definitely to be the start of the Ac_1 range in Steel N 5 of the present research. The position of the line A C at higher nickel contents is doubtful. For lower nickel contents it can be taken as very nearly correct as shown on the diagram for commercial nickel steels containing 0·5 to 0·7%.manganese. This line is obviously of considerable practical importance as it represents the temperature below which all tempering operations must be carried out.

The line A C is analagous to the solidus of a system of continuous solid solutions, and the line A E to the liquidus. A C represents the start of the Ac_1 range and A E its completion. The exact position of the line A E is doubtful. At lower nickel contents it has been drawn through Scott's figures, and it is probably at all compositions fairly near to the line obtained by joining up the Ac_1 maximums.

Consider the reactions on heating the steel N 3 (3·22% nickel). When the lower plane in Fig.26 is reached γ-solid solution of entectoid composition commences to form. The point D in Fig. 27 represents the temperature of the start of formation of the entectoid γ-solution. At this point the composition of the γ-solution first separating is given by F, i.e. the horizontal point on the line A E. The γ-solution first formed thus contains a large proportion of nickel. The carbon content of a eutectoid solution containing this amount of nickel is low, since the higher the nickel content the lower is the carbon content required to form a eutectoid solution. If the temperature is retained at this point D a condition of stability is reached through the impoverishment of the remaining α-iron in nickel, the composition of the ferrite moving in the direction shown by the arrow. It is thus necessary in order to form more eutectoid solution to raise the temperature further until the line A C is again reached. If instead of retaining the temperature at 680°C (i.e. at the point D) it is raised at once to 685°C and retained there a eutectoid solution of the approximate composition of

from F to G is formed. This contains more carbon and less nickel than in the first case when the temperature was retained at 680°C. Similarly, if the temperature is raised at once to 690°C the composition of the γ-solution extends up to H. The effect of these varying compositions on the Ar_1 change has been shown in Figs. 13, 14, and 15. The lower the soaking temperature the lower is the resulting Ar_1 point. The position of the line A E at higher nickel contents has been calculated on the assumption that increase of nickel gives a uniform decrease in the temperature of the Ar_1 point. Scott's figures (which agree closely with other workers) for low nickel steels have been taken, i.e. each 1% of nickel lowers Ar_1 by 21·5°C, and the compositions of the eutectoid γ-solutions formed at 680°C, 685°C, and 690°C were calculated from the start of the respective Ar_1 ranges (Figs. 13, 14 and 15). The points F, G, and H (Fig.27) through which the line A E has been drawn, were thus obtained. Another factor, however, may have to be taken into consideration. Dejean, (9) and Carpenter, Hadfield, and Longmuir (5) showed that above about 8 or 10% nickel the Ar_1 point is lowered rapidly to a much lower range of temperature. These workers used high initial temperatures, 900°C and higher, whereas the initial temperatures used in the present experiments were below 700°C. It is thus impossible to make a definite deduction from the available data as to the exact position of the line A E in Fig. 27 or of the corresponding plane on Fig.26. It would seem, however, that the line as shown cannot be far from its true position.

In the thermal curves on Fig.13 it is evident that sufficient time has not been given at the soaking temperature in a number of the experiments for stability to be attained. Once a eutectoid solution has been formed by soaking at a lower temperature further raising of the temperature does not readily dilute it. The soaking times used in the specific volume work are likely to have produced reasonable homogeneity of the entectoid solutions formed.

The first solutions formed on heating in the lower part of the Ac_1 range, being relatively low in carbon are more readily converted to martensite on quenching. Guillet (11) found that in quenched steels with from 7·65% to 25% nickel the microstructure showed a predominance of martensite. Evidently increase of nickel has not the same effect on the retention of austenite as increase of carbon has. Thus in the entectoid γ-solutions first formed the lower carbon content gives a tendency to formation of martensite and it is only when almost all of the carbon is brought into solution that the carbon content of the eutectoid γ-solution becomes high enough for the retention of austenite, at least when these low initial temperatures are used. Thus the dip in the specific volume curves is explained.

Some considerations on the Ar_1 range may suitably be included in this discussion, since the elucidation of the reactions which take place on heating through the A_1 range inevitably brings out information which can be applied to

throw light on the reverse process of cooling through that range.

It has been pointed out by Le Chatelier [14] and commented on by Carpenter, Hadfield, and Longmuir [5] that the Ar_1 change in steels naturally appears on a thermal curve over a wider range of temperature than Ac_1. This is due to the fact that the falling temperature opposes the Ar_1 change whereas rising temperature assists the Ac_1 change.

To this must, however, be added, in the case of steels containing nickel or other similar elements in solid solution, a factor which consideration of the ternary diagram makes clear. When nickel steels are cooled the critical changes take place at lower temperatures than they do on heating. Thus Scott [10] calculates that the Ar_1 point (maximum) is lowered 21·5°C. for each 1% of nickel, whereas the Ac_1 point is only lowered by 10·5°C. To obtain the position of the Ar_1 range the two planes in Fig. 26 must be altered in position. The point A will be slightly lowered and the planes representing the start and finish of the A_1 change turned downwards through an arc with A as centre.

A vertical section through the line of entectoid compositions has been drawn in Fig. 28. Some of the results of Scott, and of Carpenter, Hadfield, and Longmuir have been plotted on the diagram as shown. It is readily seen that on cooling there is a greater temperature interval between the two planes for any given composition than is the case on heating The Ar_1 range in steels which contain a special element in

solid solution will for this reason - among others - be greater than the Ac_1 range.

The reactions which take place in the Ar_1 range will be the reverse of those described for the Ac_1 range. The first pearlite separating will have a higher carbide content and a lower nickel content than that obtained as an average in the pearlite areas after the change is complete. Just as on heating, the first γ-iron solid solution which forms tends to absorb nickel in excess, so on cooling, when pearlite starts to separate, the nickel of the steel will tend to remain in the γ-iron solid solution which has yet to change (to pearlite, troostite, or martensite). As the nickel content of a steel rises the Ar_1 range, as shown by the two planes (represented by the two lines A E and A C, Fig 28) on the ternary diagram, widens. This, combined with the factors just discussed probably, in the opinion of the author, gives the key to the explanation of the doubling of the critical points on cooling certain nickel, nickel-chromium, and other special steels. Carpenter, Hadfield, and Longmuir [5] found in their work on nickel steels that with higher nickel contents the Ar range was revealed by a long evolution of heat culminating in a critical point near the lower extremity. Dejean [9] found in steels of similar composition that two points were revealed in certain steels. These points he named (along with other workers) Ar^1 and Ar^{11}, associated with the formation of troostite and martensite respectively. Consideration of the thermal curves published by the above workers, in the light of the new facts

brought out by the present research, points strongly to the fact that the doubling of Ar points is due to reactions which take place in the Ar_1 range. When the start of the Ar_1 range is reached on cooling, pearlite (or troostite) starts to separate and, as already pointed out, the remaining γ-iron solid solution becomes progressively richer in nickel, thus delaying its change until a lower temperature is reached. It is not difficult to imagine how two separate points may be produced in a cooling curve given suitable composition and conditions of cooling. In the present research three cooling curves of nickel-chromium steels have been shown (Figs.21, 23 and 24). In Steel N R3 (Ni.1·90%) Ar_1 is shown by a sharp point taking place over a short temperature range. In Steel N C (Ni.3·20%) the composition and conditions of cooling have led to a continuous evolution of heat throughout the Ar_1 range. In Steel A H (Ni. 4·46%) the composition and conditions have led to the production of two definite points, one at or near the start of the Ac_1 range and the other at or near the end of that range.

Other considerations complicate the reactions on cooling. After heating to higher initial temperatures the critical points of many special steels are lowered.[15] This is due to causes which are not yet understood but may be connected with the destruction of nuclei as the temperature is raised. It is sufficient here to recognise the fact and to point out how the position of the Ar_1 range will be altered with it. This, combined with the considerations already discussed, makes

clearer the fundamental reasons for the very wide ranges of Ar_1 found in many special steels, as, for example, the wide ranges found by Carpenter, Hadfield, and Longmuir at higher nickel contents.

SUMMARY

The conditions which govern the extent of the Ac_1 range in special steels and the reactions which take place in that range have been investigated by methods involving determinations of specific volume and hardness, examination of microstructures, mechanical testing, and thermal analysis, of carbon, nickel, chromium, and nickel-chromium steels.

As a result of the work the following are the principal conclusions:-

(1) Variation of carbon content of nickel steels has no effect either on the temperature at which the Ac_1 range commences, on the apparent extent of that range, or on the nature of the reactions which take place in it.

(2) Increase of nickel content of nickel and nickel-chromium steels rapidly lowers the temperature at which the Ac_1 range commences, and widens the extent of that range.

(3) At the commencement of the Ac_1 range nickel is absorbed in excess by the γ-iron eutectoid solid solution first formed, thereby impoverishing the remaining ferrite in nickel. This is evidently the fundamental physico-chemical reason for the existence of an Ac_1 range.

(4) The reactions which take place in the Ac_1 range in nickel steels can be explained by consideration of the ternary equilibrium diagram of the system iron-nickel-carbon, the necessary portion of which has been constructed for that purpose. Similar considerations will apply to other steels which contain a special element in solid solution.

(5) The elucidation of the reactions in the Ac_1 range throws light on the more complex reactions which take place in the Ar_1 range in many special steels.

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Data of Thesis

by

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PART I

TABLE 1

Specimens O.H. 850°C. tempered 3 Hours at each temperature and quenched in water.

Steel	450°C.		500°C.		550°C.	
	Sp.Vol.	B.H.	Sp. Vol.	B.H.	Sp. Vol.	B.H.
A22	0·127401	258	0·127410	219	0·127364	201
NP 1	0·127407	298	0·127391	273	0·127384	241
NP 5	0·127402	302	0·127400	273	0·127396	247
AW22	0·127548	365	0·127511	342	0·127506	309
AW23	0·127666	369	0·127640	326	0·127653	302

Steel.	600°C.		650°C.		670°C.		680°C.	
	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.
A22	0·127359	194	0·127365	175	0·127370	174	0·127379	160
NP 1	0·127370	209	0·127312	197	0·127224	194	0·127158	208
NP5	0·127397	225	0·127334	209	0·127214	207	0·127180	[illegible]
AW22	0·127491	286	0·127485	256	0·127449	253	0·127392	256
AW23	0·127619	285	0·127616	262	0·127602	248	0·127593	247

Steel.	690°C.		695°C.		700°C.		705°C.	
	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.
A22	0·127365	166	0·127380	155	0·127375	185	0·127392	161
NP 1	0·127178	239	0·127192	240	0·127185	207	0·127394	273
NP 5	0·127317	278	0·127328	270	0·127287	258	0·127571	328
AW22	0,127380	253	0·127354	256	0·127420	266	0·127549	305
AW23	0·127552	253	0·127558	254	0·127657	276	0·127654	276

Steel	710°C.	
	Sp.Vol.	B.H.
A22	0·127392	151
NP 1	0·127543	326
NP 5	0·127628	341
AW22	0·127628	326
AW23	0·127666	280

TABLE 1. (Continued)

Specimens O.H. 850°C. tempered 3 hours at each temperature and slowly cooled at 2·0°C. per minute.

Steel	450°C. Sp.Vol.	B.H.	500°C. Sp.Vol.	B.H.	550°C. Sp.Vol.	B.H.	600°C. Sp.Vol.	B.H.
A22	0·127374	256	0·127386	241	0·127352	220	0·127378	196
NP 1	0·127387	299	0·127399	263	0·127358	237	0·127383	208
NP 5	0·127407	300	0·127411	273	0·127364	242	0·127401	222
AW22	0·127543	364	0·127531	342	0·127474	314	0·127484	278
AW23	0·127674	367	0·127614	321	0·127568	243	0·127543	228

Steel	650°C. Sp.Vol.	B.H.	670°C. Sp.Vol.	B.H.	680°C. Sp.Vol.	B.H.	690°C. Sp.Vol.	B.H.
A22	0·127369	184	0·127346	170	0·127322	172	0·127358	182
NP 1	0·127338	194	0·127248	194	0·127251	190	0·127127	194
NP 5	0·127327	201	0·127285	198	0·127250	212	0·127195	215
AW22	0·127450	252	0·127423	242	0·127410	239	0·127281	239
AW23	0·127488	212	0·127388	209	0·127403	210	0·127418	212

Steel	695°C. Sp.Vol.	B.H.	700°C. Sp.Vol.	B.H.	705°C. Sp.Vol.	B.H.	710°C. Sp.Vol.	B.H.
A22	0·127376	188	0·127371	151	0·127364	170	0·127374	170
NP 1	0·127127	194	0·127129	200	0·127204	194	0·127216	201
NP 5	0·127160	213	0·127185	230	0·127178	222	0·127180	215
AW22	0·127265	238	0·127344	240	0·127326	256	0·127370	256
AW23	0·127450	240	0·127495	240	0·127527	248	0·127546	256

TABLE 2

Steel	Original Treatment. Tempered three Hours at 450°C.			Repeat at end of Experiments. Tempered three Hours at 450°C.		
		Specific Volume	Brinell Hardness		Specific Volume	Brinell Hardness
A22	Quenched	0·127401	258	Quenched	0·127381	258
	Slowly cooled	0.127374	256	"	0.127418	256
NP 1	Quenched	0.127407	298	"	0.127397	289
	Slowly cooled	0.127387	299	"	0.127416	287
NP5	Quenched	0.127430	302	"	0.127448	300
	Slowly cooled	0.127407	300	"	0.127432	298
AW22	Quenched	0.127548	365	"	0.127551	356
	Slowly cooled	0.127543	364	"	0.127552	361
AW23	Quenched	0.127666	369	"	0.127679	365
	Slowly cooled	0.127677	367	"	0.127678	367

Steel	Original Treatment. Tempered three Hours at 670°C.			Repeat Treatment. Tempered six Hours at 670°C.		
		Specific Volume	Brinell Hardness		Specific Volume	Brinell Hardness
AW23	Quenched	0·127602	248	Quenched	0·127580	240
	Slowly cooled	0·127388	209	"	0·127615	240

TABLE 3.

Specimens continuously tempered at 650°C.

Quenched in water at end of each period shown.

Total Soaking Time

Steel	2 hours Sp.Vol.	B.H.	8 hours Sp.Vol.	B.H.	14 hours Sp.Vol.	B.H.	20 hours Sp.Vol.	B.H.
AH	0·127706	270	0·127684	250	0·127691	235	0·127642	230
NC	0·127582	256	0·127543	232	0·127550	224	0·127551	218
NR3	0·127636	256	0·127596	241	0·127598	232	0·127602	228
AW22	0·127493	257	0·127469	239	0·127466	239	0·127436	228
AW23	0·127640	260	0·127617	240	0·127622	230	0·127608	224

Total Soaking Time

Steel	26 hours Sp.Vol.	B.H.	32 hours Sp.Vol.	B.H.	38 hours Sp.Vol.	B.H.
AH	0·127649	227	0·127657	222	0·127665	222
NC	0·127542	209	0·127541	204	0·127544	207
NR3	0·127586	217	0·127602	218	0·127550	217
AW22	0·127441	224	0·127441	222	0·127417	222
AW23	0·127622	217	0·127603	217	0·127590	218

Total Soaking Time

Steel	44 hours Sp.Vol.	B.H.	50 hours Sp.Vol.	B.H.	56 hours Sp.Vol.	B.H.
AH	0·127687	218	0·127690	214	0·127685	220
NC	0·127539	207	0·127564	201	0·127538	203
NR3	0·127585	217	0·127592	208	0·127603	211
AW22	0·127451	217	0·127453	214	0·127462	217
AW23	0·127611	214	0·127607	207	0·127629	211

TABLE 4.

Tempering Temperatures.

Steel	650°C. Quenched Izod	650°C. Quenched B.H.	650°C. Slow-cooled Izod	650°C. Slow-cooled B.H.	670°C. Quenched Izod.	670°C. Quenched B.H.	670°C. Slow-cooled Izod	670°C. Slow-cooled B.H.
AH	58·7	254	5·5	274	34·4	297	7·7	268
NC	61·8	258	5·1	252	73·2	251	6·1	283
NR3	69·5	253	8·3	256	79·5	238	15·0	234

Steel	680°C. Quenched Izod	680°C. Quenched B.H.	680°C. Slow-cooled Izod	680°C. Slow-cooled B.H.	690°C. Quenched Izod.	690°C. Quenched B.H.	690°C. Slow-cooled Izod	690°C. Slow-cooled B.H.
AH	27.2	308	5.2	260	8.6	346	5·2	272
NC	66·6	237	5·6	225	17·1	309	7·0	283
NR3	78·7	236	10·9	237	68·4	220	10·2	251

TABLE 5

Steel	(1) Oil-hardened & Tempered 650°C. for two hours and Water-quenched.		(2) Oil-hardened & Tempered 650°C. for two hours & cooled at 0·3°C. per minute.		(3) Oil-hardened and Tempered 670°C. for 2½ hours and cooled at 0·3°C. per minute.		(4) (3) Retempered 550°C. for three hours and cooled at 0·3°C. per minute.		(5) Oil-hardened & Tempered 670°C. for three hours and cooled at 2°C. per minute.	
	Specific Volume.	Brinell Hardness.	Specific Volume.	Brinell Hardness.	Specific Volume.	Brinell Hardness.	Specific Volume.	Brinell Hardness	Specific Volume.	Brinell Hardness.
AH	0·127694	278	0·127698	267	0·127693	266	0·127716	266	0·127577	[illegible]
HC	0·127556	264	0·127617	242	0·127559	256	0·127565	254	0·127524	[illegible]
MA3	0·127611	266	0·127602	263	0·127630	269	0·127596	270	0·127543	[illegible]
AM22	0·127490	257	0·127466	247	0·127481	252	0·127487	249	0·127337	[illegible]
AM33	0·127619	265	0·127592	230	0·127591	238	0·127613	235	0·127492 0·127482 0·127490	[illegible]
A22	0·127364	177	0·127368	170	0·127361	168	0·127361	170	0·127351	165
NP 1	0·127330	198	0·127363	195	0·127296	192	0·127318	190	0·127153	[illegible]
NP5	0·127353	211	0·127288	201	0·127264	196	0·127303	196	0·127249	[illegible]
AAX	0·127625	261	0·127569	257	0·127631	215	0·127650	215	0·127620	[illegible]
[illegible]	...	...	0·128180	401	0·128108	384	0·128075	381	0·128014	[illegible]

TABLE 6

Steel	Quenched		Cooling Rate 3°C./min.		2°C./min.		1°C./min.		0·6°C./min.	
	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.
A22	0·127364	177	0·127371	174	0·127370	175	0·127368	175	0·127369	173
AH	0,127694	278	0·127620	277	0·127646	265	0·127673	265	0·127683	261
NC	0·127556	264	0·127576	255	0·127554	256	0·127519	262	0·127517	258
NR 3	0·127611	266	0·127608	252	0·127595	264	0·127589	264	0·127581	263
N3	0·127490	243	0·127482	235	0·127478	238	0·127475	240	0·127456	240
NP5	0·127353	211	0·127300	201	0·127336	208	0·127357	209	0·127354	210
AAX	0·127625	261	0·127612	249	0·127635	248	0·127602	252	0·127600	257

TABLE 7.

Steel	6 hours 300°C.		6 hours. 400°C.		6 hours 450°C.		6 hours 500°C.	
	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.
AH	0·127942	408	0·127801	385	0·127760	358	0·127763	330
NC	0·127801	415	0·127732	391	0·127692	360	0·127673	325
NR3	0·127714	391	0·127651	377	0·127639	356	0·127636	320
AW22	0·127625	380	0·127556	361	0·127538	354	0·127524	300
AW23	0·127762	400	0·127728	390	0·127696	365	0·127682	333

Steel	6 hours 550°C.		6 hours 600°C.		6 hours 650°C.		3 hours 675°C.	
	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.	Sp.Vol.	B.H.
AH	0·127762	304	0·127751	283	0·127748	254	0·127695	243
NC	0·127646	296	0·127622	270	0·127623	245	0·127589	226
NR3	0·127597	300	0·127581	278	0·127575	237	0·127552	228
AW22	0·127503	275	0·127484	266	0·127481	247	0·127420	224
AW23	0·127638	291	0·127632	276	0·127609	248	0·127585	224

Steel	Extra 3 hours 675°C.	
	Sp.Vol.	B.H.
AH	0·127666	231
NC	0·127585	224
NR3	0·127544	221
AW22	0·127406	214
AW23	0·127578	220

TABLE 8.

Experiments on Slowly-Cooled Nickel Steel with Grain Boundary

Experiment (b)

Specimen	(Grain boundary Condition) Specific Volume.	Brinell Hardness	Microscopical Observations.
1. (from 1st break in Izod.bar)	0·127427	220	Distinct grain boundary but more Carbide throughout mass than in others.
2. (from 2nd break in Izod bar)	0·127479	226	Grain boundary more than No. 1.
3. (from 3rd break in Izod bar)	0·127456	230	Grain boundary very obvious.
4. (from original micro-specimen)	0·127458	243	Grain boundary plainer than in any of the others.

The Grain boundary increases from 1 to 4 and the amount of carbide in the grains varies inversely.

Difference between maximum and minimum B.H. = 23

Experiment (c)

Tempered Fifteen minutes at 650°C.

Specimen	Specific Volume.	Brinell Hardness	Microscopical Observations
1.	0 127372	177	Grain boundary largely broken up into small globules but still showing a good deal round grains. Carbide throughout mass unchanged in amount but globularised a little. Grain size unchanged.
2.	0 127389	191	
3.	0 127395	194	
4.	0 127363	183	

Difference between maximum and minimum B.H. = 17

TABLE 9.

Experiment 1.

O.H. 850°C., heated to 760°C. and slowly cooled through Ar_1 at 1°C. per minute.

Specimen	Specific Volume	Brinell Hardness	Microscopical Observations
2. (see Table 8)	0·127345	165	No boundary. Carbide throughout mass.
3. (see Table 8)	0·127332	171	
Al_2 (C,0·22%)	0·127400	113	Obvious grain boundary, but much broken up.
$A22_1$ (C,0·48%)	0·127331	148	No grain boundary.

Experiment 2.

O.H. 900°C., heated to 730°C. (above Ac_1) and cooled in furnace. Rate of cooling through Ar_1 = 40°C. per minute.

Specimen	Specific Volume.	Brinell Hardness	Microscopical Observations
Al_1 (C,0·22%)	0·127423	121	Tendency for carbide to be round grain.
Al_2 (C,0·22%)	0·127415	121	
$A22_1$ (C,0·48%)	0·127349	167	Mixture of ferrite and pearlite.
$A22_2$ (C,0·48%)	0·127367	173	

TABLE 9.(continued)

Experiment 3.

O.H. 900°C., W.T. 1 hour 650°C.

Specimen	Specific Volume.	Brinell Hardness	Microscopical Observations.
1. (see Table 8)	0·127474	233	
N3	0·127480	238	Not examined
Al_2 (C,0·22%)	0·127432	142	
$A22_1$ (C,0·48%)	0·127378	192	

Experiment 4.

O.H. 850°C., reheated to 760°C. and slowly cooled at 1°C. through whole range to below Ar_1.

Specimen	Specific Volume.	Brinell Hardness	Microscopical Observations.
2.(see Table 8)	0·127309	171	Grain size very small.
3.(see Table 8)	0·127241	176	No boundary. Carbide
N3.	0·127404	176	in globules.

TABLE 9 (continued)

Experiment 5.

Reheated to 760°C. for two hours and cooled at 1°C. per minute over whole range to below Ar_1

Specimen (see Experiment 4)	Brinell Hardness	Microscopical Observations
2.	171	No boundary, small grains.
3.	171	
N3.	177	

Experiment 6.

Reheated to 760°C. for 2 hours and cooled at 0·6°C. per minute to below Ar_1.

Specimen	Brinell Hardness	Microscopical Observations
2.	163	No difference from Experiment 5.
3.	170	
N3.	165	

TABLE 10

Experiments on Nickel Steel quenched Izod Bar,

Specimen	Experiment 1. Original Condition.		Experiment 2. Heated to 590°C. and slow cooled to 430°C.		Experiment 3. Reheated to 430°C. for 15 hours.		Experiment 4. Heated to 760°C. and cooled in furnace.	
	Sp.Vol.	B.H.	Sp.vol.	B.H.	Sp.vol.	B.H.	Sp.Vol.	B.H.
1. (from 1st Izod fracture)	0·127385	283	0·127365	214	0·127368	217	0·127353	206
2. (from 2nd Izod fracture)	0·127353	273	0·127436	216	0·127436	214	not treated.	
3. (from 3rd Izod fracture)	0·127368	272	0·127455	204	0·127451	207	0·127324	200
4. (from original micro-specimen)	0·127409	285	0·127430	198	0·127441	197	0·127290	200

Microstructures examined after Experiment 3, showed finely divided carbide; no boundary; no graphite.

TABLE 11.

Specific Resistance, ρ
at 20°C.
in microhms/c.c.

Steel	Condition (1) Tough	(2) Intermediate	(3) Brittle	Percentage drop in ρ (1) - (2)	(2) - (3)	(1) - (3)
AH	32·20	31·62	31·21	1·8	1·3	3·17
NC	24·23	23·94	23·69	1·2	1·05	2·22
NR3	30·47	30·10	29·92	1·2	0·6	1·8

Analysis of Steels

	C.	Mn.	Si.	P.	S.	Ni.	Cr.
AH	0·31	0·47	0·310	0·015	0·021	4·46	1·41
NC	0·31	0·57	0·145	0·026	0·029	3·20	0·83
NR3	0·36	0·34	0·275	0·016	0·021	1·90	1·15

TABLE 12

Tempering Temperature		Steel A H		Steel N C		Steel N R 3	
		Specific Resistance	% Drop	Specific Resistance	% Drop	Specific Resistance	% Drop
650°C	Quenched	32.61	3·43	24.78	2·9	26.51	3·05
	Slow Cooled	31.49		24.06		25.70	
670°C	Quenched	33·64	5.1	24·54	2·44	26·53	2·41
	Slow Cooled	31·92		23·94		25·89	
680°C	Quenched	33·82	9·16	24·70	2·92	26·29	1·98
	Slow Cooled	30·72		23·98		25·67	
690°C	Quenched	38·83	16·0	25·27	3·83	29·19	12·2
	Slow Cooled	32·59		24·30		25·64	

TABLE 13.

Steel A H. Tough Condition.

Final Treatment:- tempered 2 hours 660°C. and waterquenched.

(a) Hysteresis Loop for H max = 150

H	B descending.	B ascending.
150	17,550	17,550
100	16,950	16,730
50	15, 980	14,970
30	15,360	12,170
25	-	10,030
20	14,970	5,630
18	-	3,150
16	-	0
0	13,320	-
-10	10,640	-
-12	9,220	-
-14	4,500	-
-15	2,100	-
-16	0	-

Remanence = 13,320. Coercive force = 16·0.

(b) Variation of B max and B rem with H.

H	B max	μ	B rem.
10	1,500	150	450
14	4,600	329	2,750
20	8,500	425	6,320
30	12,400	413	9,910
50	14,960	299	11,930
70	15,940	228	12,670
100	16,780	168	13,110
150	17,550	117	13,320
200	18,070	90	13,340
300	18,710	62	13,340
400	19,100	48	13,340

TABLE 14.

Steel A H. Intermediate Condition.

Final Treatment:- Tempered 2 hours 660°C., and slow cooled at 2·0°C. / min.

(a)

Hysteresis Loop

H	B desc.	B asc.
150	17,380	17,380
100	16,700	16,500
50	15,780	14,800
30	15,300	13,200
25	-	12,380
20	14,910	10,400
18	-	5,250
15·2	-	0
0	13,525	-
-10	11,650	-
-12	9,010	-
-14	4,200	-
-15·2	0	-

Remanence = 13,525. Coercive force = 15·2

(b)

H	B max	μ	B rem
10	1,700	170	500
13	4,500	346	2,000
20	10,500	525	7,900
30	13,200	440	10,300
50	14,820	296	11,900
70	15,700	224	12,750
100	16,500	165	13,300
150	17,380	116	13,525
200	17,800	89	13,580
250	18,200	73	13,580

TABLE 15.

Steel A H. Brittle Condition

Final Treatment:- Tempered 2 hours 660°C., and slow cooled at 0·3°C./min.

(a)

Hysteresis Loop.

H	B desc.	B asc.
150	17,460	17,460
100	16,900	16,600
50	16,180	14,950
30	15,610	13,600
25	-	12,750
20	15,300	11,300
17	-	9,000
15	-	5,000
13·7	-	0
0	14,170	-
-10	11,200	-
-12	5,000	-
-13·7	0	-

Remanence = 14,170. Coercive force = 13·7

(b)

H.	B max.	μ.	B rem.
7	1,000	143	300
10	5,500	550	1,000
13	8,500	654	5,000
20	11,700	585	10,100
30	13,600	453	11,700
50	14,950	299	12,900
100	16,640	166	13,900
150	17,460	116	14,170
200	17,810	89	14,200
250	18,100	73	14,200

TABLE 16.

Steel N C. Tough Condition.

Final Treatment:- Tempered 2 hours 660°C., and water quenched.

(a)

Hysteresis Loop.

H.	B desc.	B asc.
150	18,440	18,440
100	17,700	17,500
50	16,600	15,300
30	15,900	13,700
20	15,490	11,500
17	-	7,500
14·1	-	0
0	14,030	-
-10	10,750	-
-12	7,000	-
-14·1	0	-

Remanence = 14,030. Coercive force = 14·1

(b)

H.	B max	μ	B rem
10	3,400	340	480
15	8,500	566	6,000
20	11,500	575	9,100
30	13,750	458	11,400
50	15,400	308	12,800
70	16,380	234	13,550
100	17,490	175	13,850
150	18,440	122	14,030
200	18,700	94	14,100
250	18,900	76	14,100

TABLE 17.

Steel N. C. Intermediate Condition.

Final Treatment:- Tempered 2 hours 660°C., and slow cooled at 2·0°C/Min.

(a)

Hysteresis Loop.

H.	B Desc.	B asc.
150	18,350	18,350
100	18,000	17,650
70	17,600	16,950
50	17,250	16,200
20	16,500	13,550
14	-	11,000
12	-	6,500
0	15,400	-
-10	11,500	-
-11	6,500	-
-12·6	0	-

Remanence = 15,400 Coercive force = 12 6

(b)

H.	B max.	μ.	B rem.
8	3,000	375	-
10	8,200	820	1,500
14	11,250	800	8,000
20	13,500	675	11,050
50	16,200	324	14,000
70	16,940	242	14,650
100	17,650	176	15,150
150	18,350	122	15,400
200	18,500	93	15,600
250	18,750	75	15,600.

TABLE 18.

Steel N C. Brittle Condition.

Final Treatment:- Tempered 2 hours 660°C, and slow cooled at 0·3°C/Min.

(a)

Hysteresis Loop.

H.	B Desc.	B asc.
150	18,210	18,210
100	17,850	17,650
70	17,500	16,900
40	17,000	15,750
20	16,510	10,000
15	-	8,700
13	-	4,000
11·8	-	0
0	15,480	-
-8	12,500	-
-10	8,500	-
-11·8	0	-

Remanence = 15,480. Coercive force = 11·8.

(b)

H.	B max.	μ.	B rem.
8.	3,500	117	-
10	8,000	800	1,000
14	11,500	825	7,000
20	13,600	680	11,000
30	15,100	503	12,800
50	16,200	324	14,000
70	16,900	241	14,600
100	17,610	176	15,150
150	18,210	121	15,480
200	18,500	93	15,650
250	18,730	75	15,660.

TABLE 19.

Steel N R3 Tough Condition.

Final Treatment:- Tempered 2 hours 660°C., and water quenched.

(a)

Hysteresis Loop

H.	B dex.	B asx.
150	18,450	18,450
100	17,550	17,250
70	16,750	16,050
40	15,820	14,400
20	15,000	11,750
15	-	8,500
13	-	4,500
12·2	-	0
0	13,420	-
-10	9,500	-
-11	4,500	-
-12·2	0	-

Remanence = 13,420. Coercive force = 12·2.

(b)

H.	B max.	μ.	B rem.
5	1,000	200	-
8	4,000	500	2,000
13	8,100	624	6,000
20	11,700	585	9,500
40	14,500	362	11,800
70	16,050	229	12,750
100	17,250	173	13,150
150	18,450	123	13,420
200	19,000	95	13,500
250	19,300	77	13,500

TABLE 20.

Steel N R 3. Intermediate Condition.

Final Treatment:- Tempered 2 hours 660° C., and slow cooled at 2·0 °C./min.

(a)

Hysteresis Loop.

H.	B desc.	B asc.
150	18,000	18,000
100	17,500	17,200
70	17,000	16,300
40	16,300	14,800
25	-	13,500
20	15,600	12,000
18	-	9,000
14	-	5,000
12·6	-	0
0	14,400	-
-10	10,000	-
-11	5,020	-
-12·6	0	-

Remanence = 14,400. Coercive force = 12·6.

(b)

H.	B max.	μ.	B rem.
5	1,250	250	-
8	4,000	500	1,100
10	6,200	620	3,450
14	9,400	670	6,500
20	11,950	598	10,000
25	13,450	538	11,000
40	14,800	370	12,500
70	16,300	233	13,650
100	17,200	172	14,180
150	18,000	120	14,400
200	18,500	93	14,600
250	18,800	75	14,600

TABLE 21.

Steel N R 3. Brittle Condition.

Final Treatment:- Tempered 2 hours 660°C., and slow cooled at 0·3°C./min.

(a)

Hysteresis Loop.

H.	B desc.	B asc.
150	18,410	18,410
100	17,850	17,610
70	17,450	16,700
40	16,750	15,300
25	-	14,000
20	16,000	12,500
17	-	9,520
13	-	5,100
11·8	-	0
0	14,870	-
-8	11,500	-
-10	6,000	-
-11·8	0	-

Remanence = 14,870. Coercive force = 11·8.

(b)

H.	B max.	μ.	B rem.
7	1,500	214	-
8	4,500	562	1,000
10	8,000	800	3,800
14	10,400	745	8,500
20	12,600	630	10,400
30	14,700	490	12,100
50	15,950	319	13,500
70	16,700	239	14,100
100	17,600	176	14,580
150	18,410	123	14,870
200	18,700	93	15,000
250	19,000	76	15,000

TABLE 22

Characteristics of Hysteresis Loops for H max. = 150

AH	Coercive Force Hc.	Maximum Induction Bm.	Remanent Induction Br.	Hysteresis Area. Hc x Br	Percentage Rise in Remanence.	
AH (tough)	16·0	17,550	13,320	$2·130 \times 10^5$		
					1·54	
AH (inter)	15·2	17,380	13,525	$2·055 \times 10^5$		6·38
					4·84	
AH (brittle)	13·7	17,460	14,170	$1·943 \times 10^5$		
NC (tough)	14·1	18,440	14,030	$1·978 \times 10^5$		
					9·77	
NC (inter)	12·6	18,350	15,400	$1·940 \times 10^5$		10·33
NC (brittle)	11·8	18,210	15,480	$1·828 \times 10^5$	0·52	
NR3 (tough)	12.2	18,450	13,420	1.635×10^5		
					7·32	
NR3 (inter)	12·6	18,000	14,400	$1·815 \times 10^5$		10·80
					3·263	
NR3 (brittle)	11·8	18,410	14,870	$1·754 \times 10^5$		

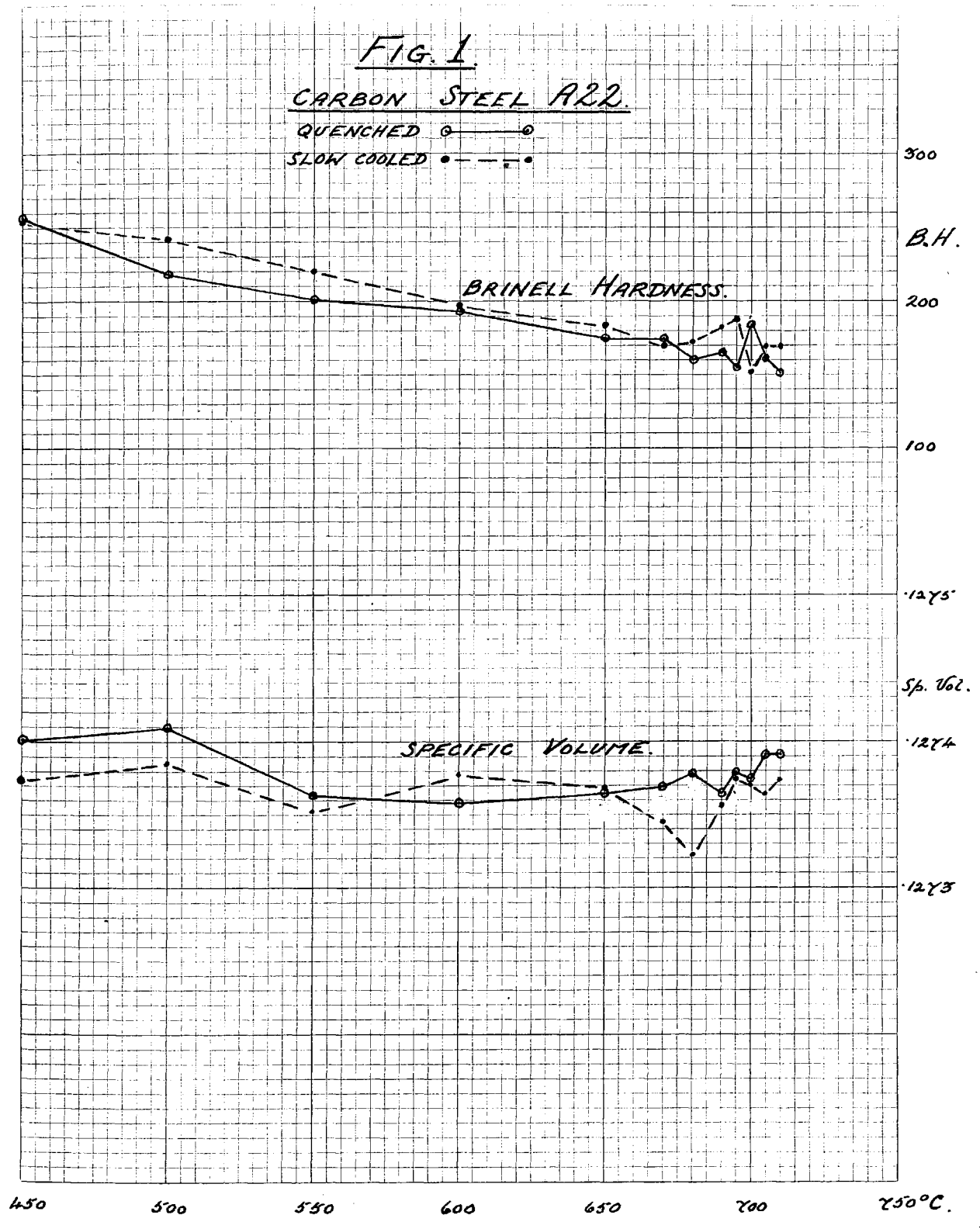


FIG. 1
CARBON STEEL A22.
QUENCHED
SLOW COOLED
BRINELL HARDNESS.
SPECIFIC VOLUME.
300
B.H.
200
100
·1275
Sp. Vol.
·1274
·1273
450
500
550
600
650
700
750°C.
TEMPERING TEMPERATURE.

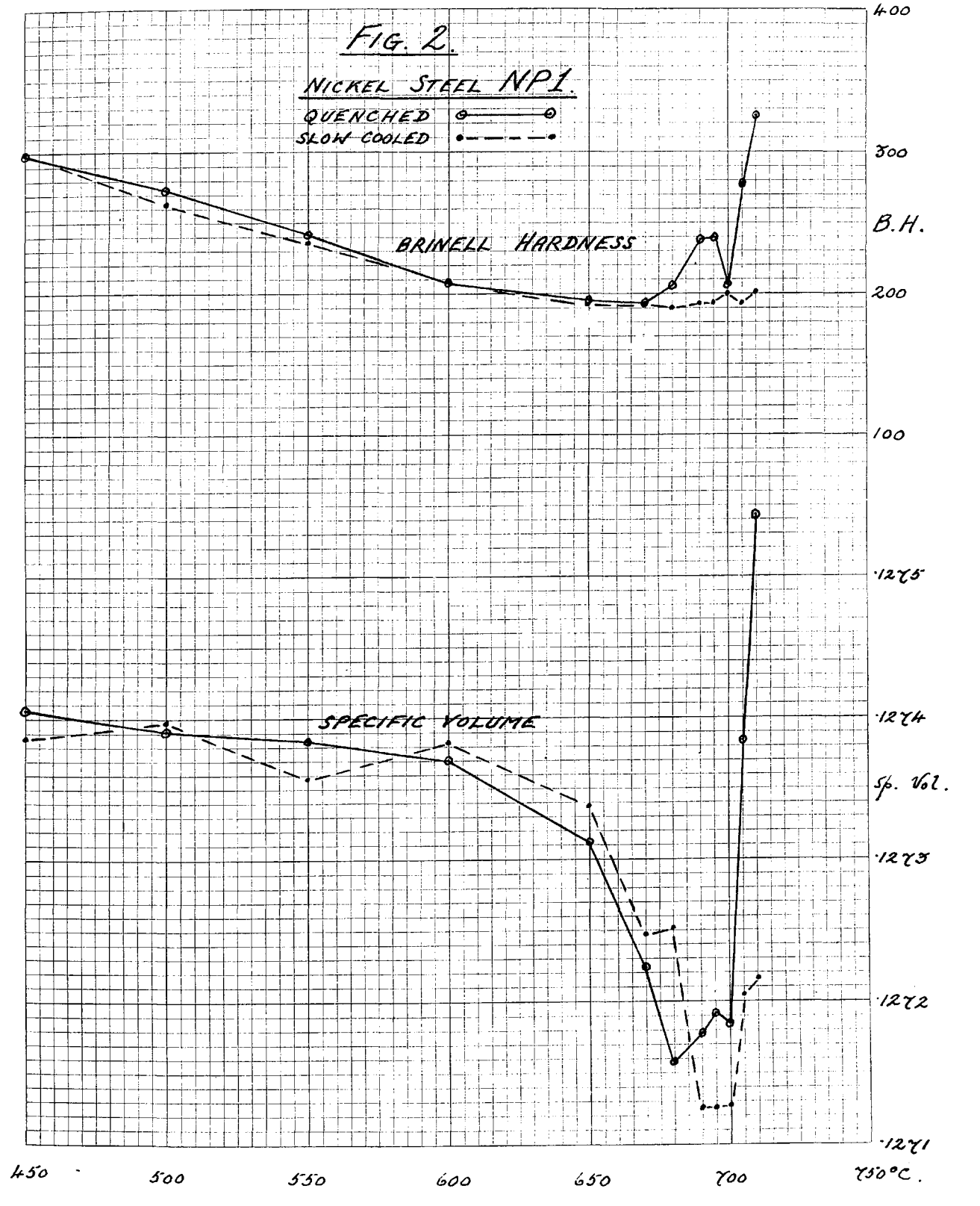


FIG. 2.
NICKEL STEEL NP1.
QUENCHED
SLOW COOLED
BRINELL HARDNESS
B.H.
400
300
200
100
SPECIFIC VOLUME
Sp. Vol.
·1275
·1274
·1273
·1272
·1271
450
500
550
600
650
700
750°C.
TEMPERATURE.

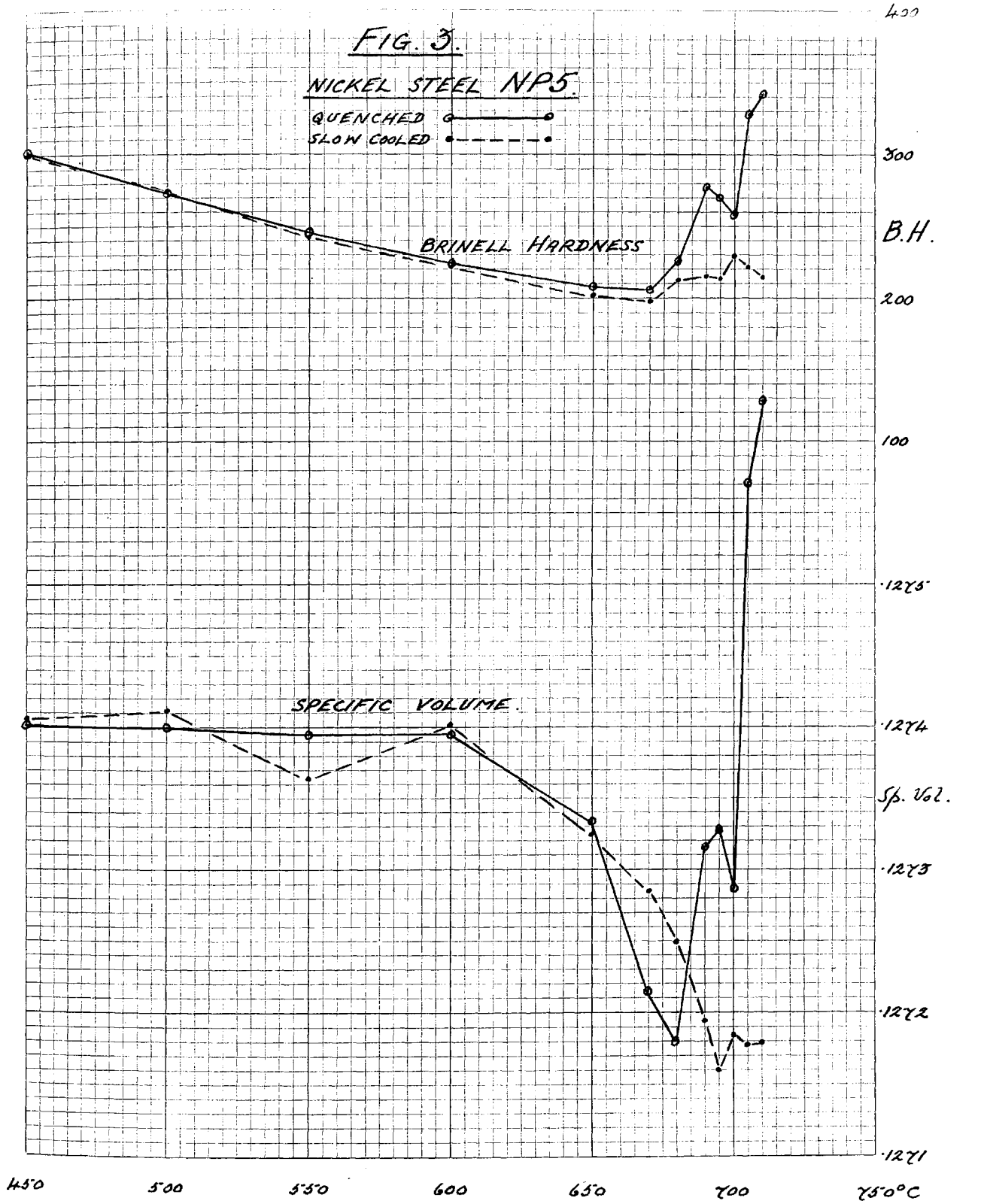


FIG. 3.
NICKEL STEEL NP5.
QUENCHED
SLOW COOLED
BRINELL HARDNESS
SPECIFIC VOLUME.
400
300
B.H.
200
100
·1275
·1274
Sp. Vol.
·1273
·1272
·1271
450
500
550
600
650
700
750°C
TEMPERING TEMPERATURE.

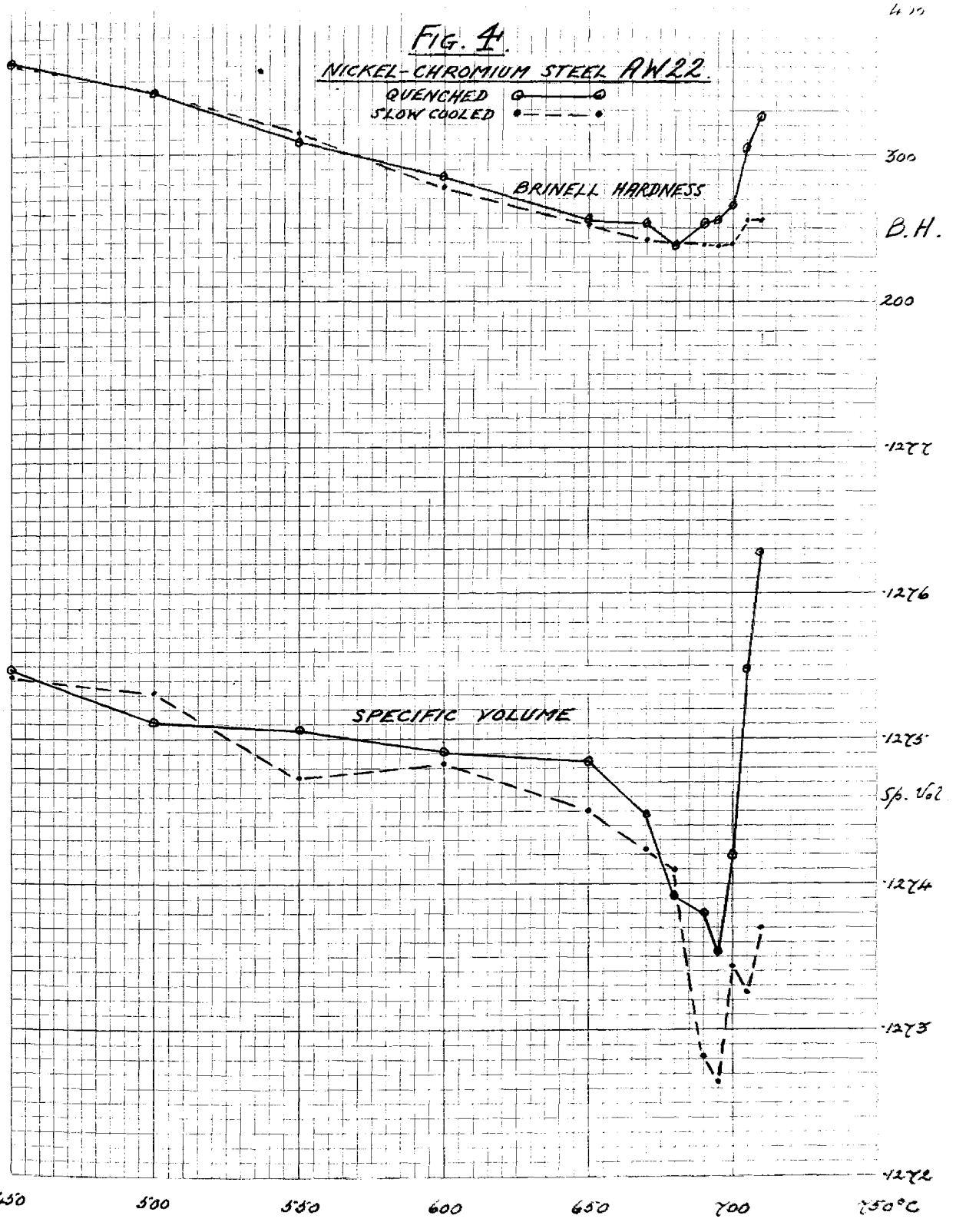


FIG. 4.
NICKEL-CHROMIUM STEEL AW22.
QUENCHED
SLOW COOLED
BRINELL HARDNESS
300
B.H.
200
·1277
·1276
SPECIFIC VOLUME
·1275
Sp. Vol
·1274
·1273
·1272
450
500
550
600
650
700
750°C
TEMPERING TEMPERATURE.

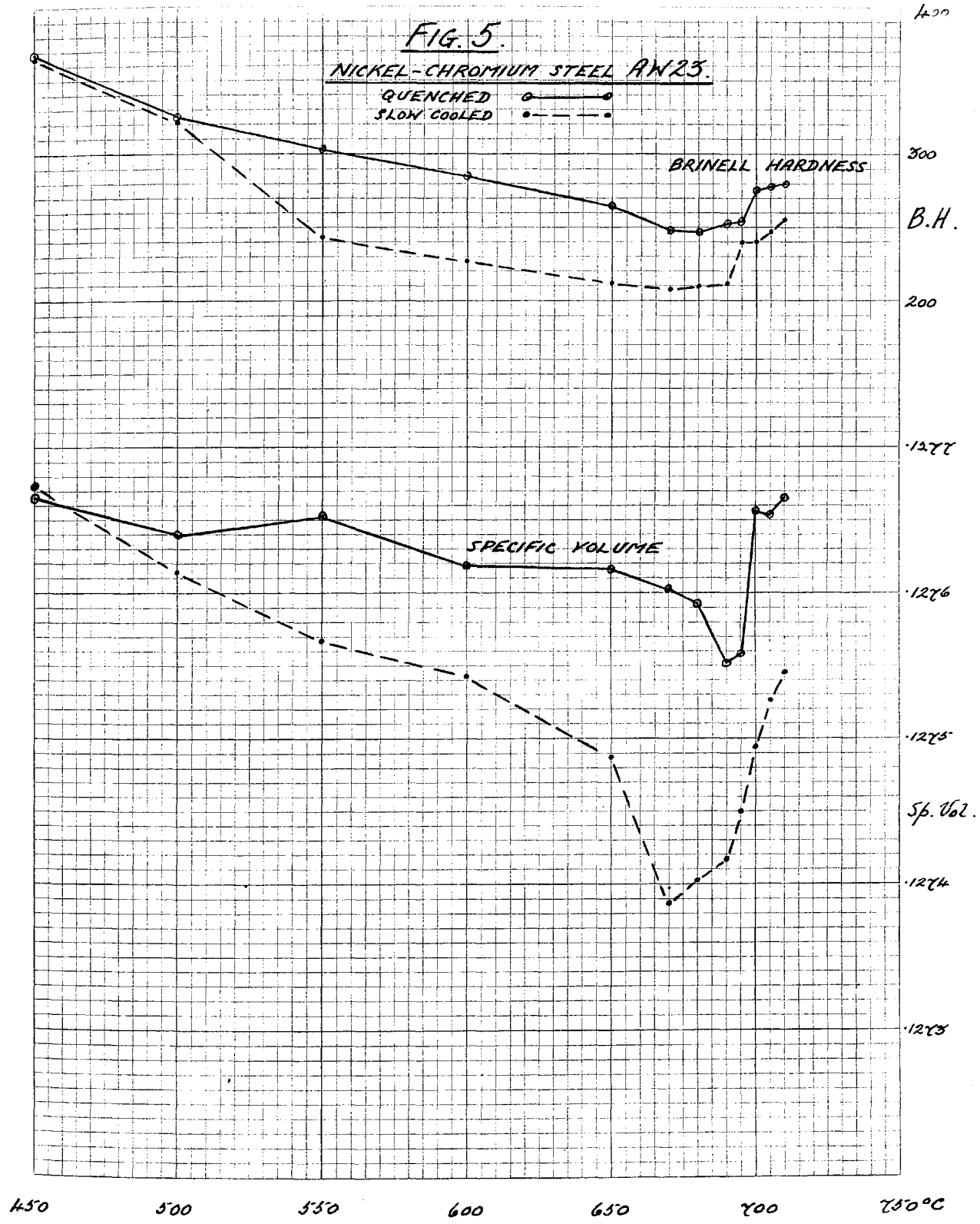


FIG. 5.
NICKEL-CHROMIUM STEEL AN23.
QUENCHED
SLOW COOLED
BRINELL HARDNESS
B.H.
SPECIFIC VOLUME
Sp. Vol.
400
300
200
·1277
·1276
·1275
·1274
·1273
450
500
550
600
650
700
750°C
TEMPERING TEMPERATURE.

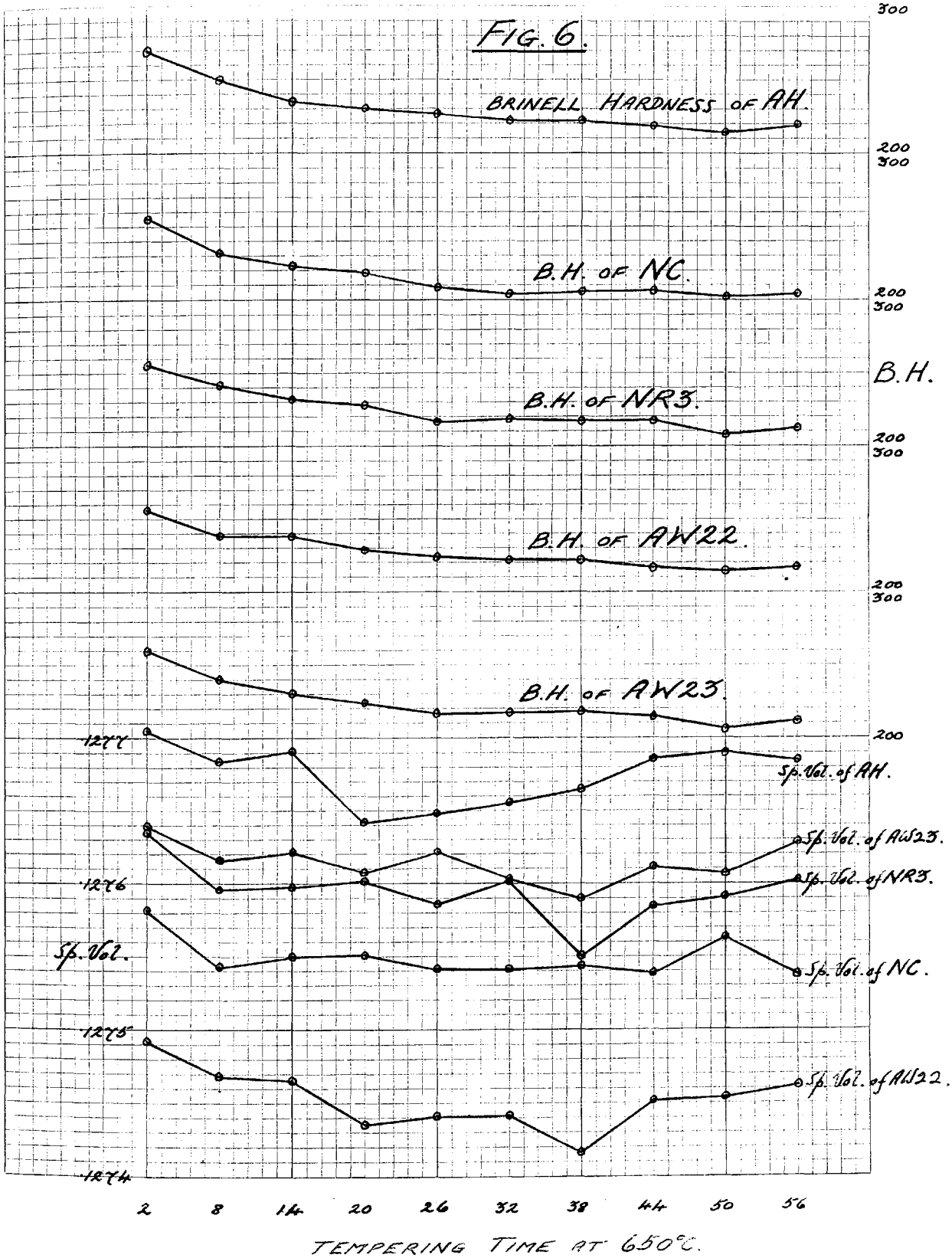


FIG. 6.
BRINELL HARDNESS OF AH.
B.H. OF NC.
B.H. OF NR3.
B.H. OF AW22.
B.H. OF AW23.
300
200
300
200
300
200
300
200
300
200
B.H.
1277
1276
1275
1274
Sp. Vol.
Sp. Vol. of AH.
Sp. Vol. of AW23.
Sp. Vol. of NR3.
Sp. Vol. of NC.
Sp. Vol. of AW22.
2
8
14
20
26
32
38
44
50
56
TEMPERING TIME AT 650°C.

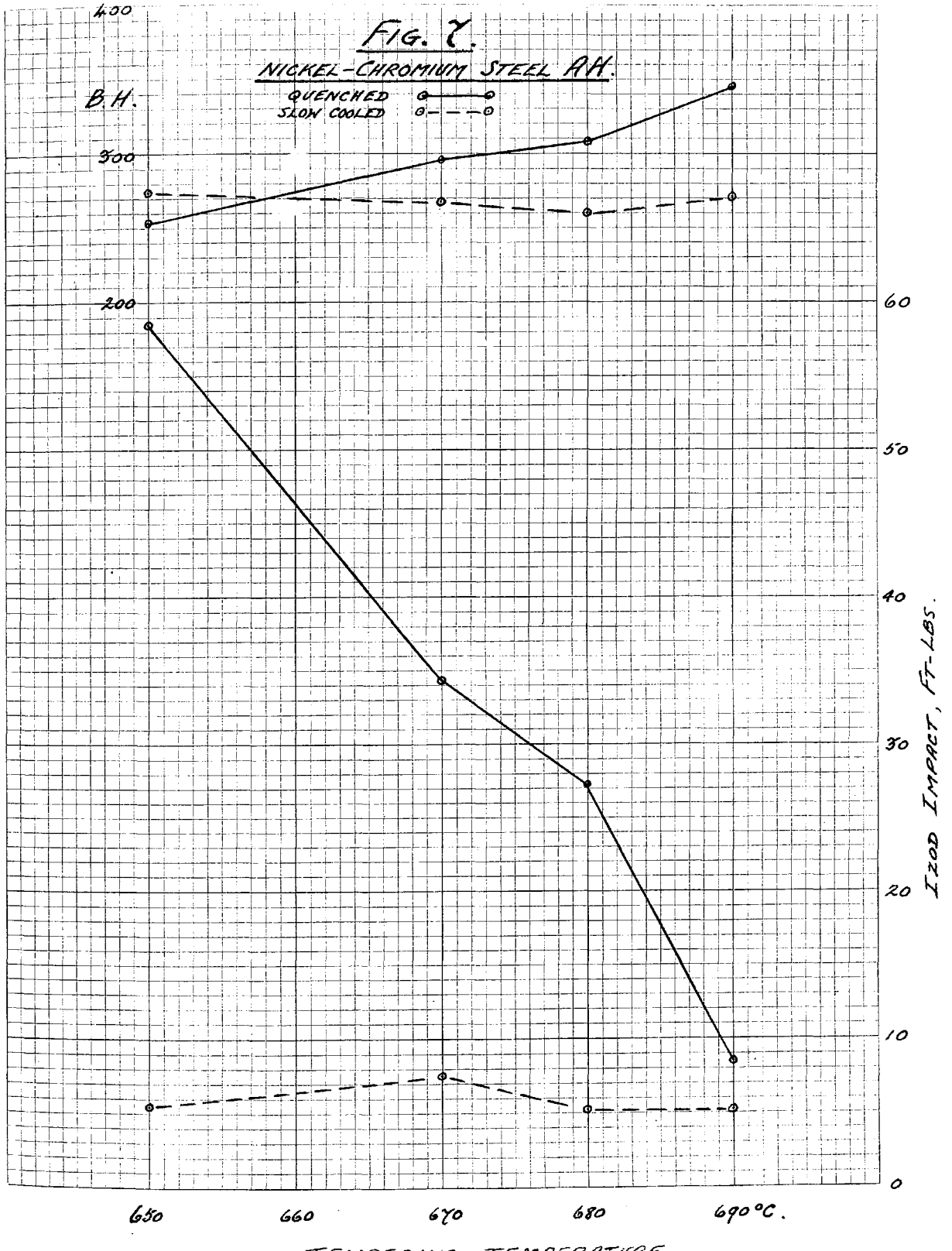
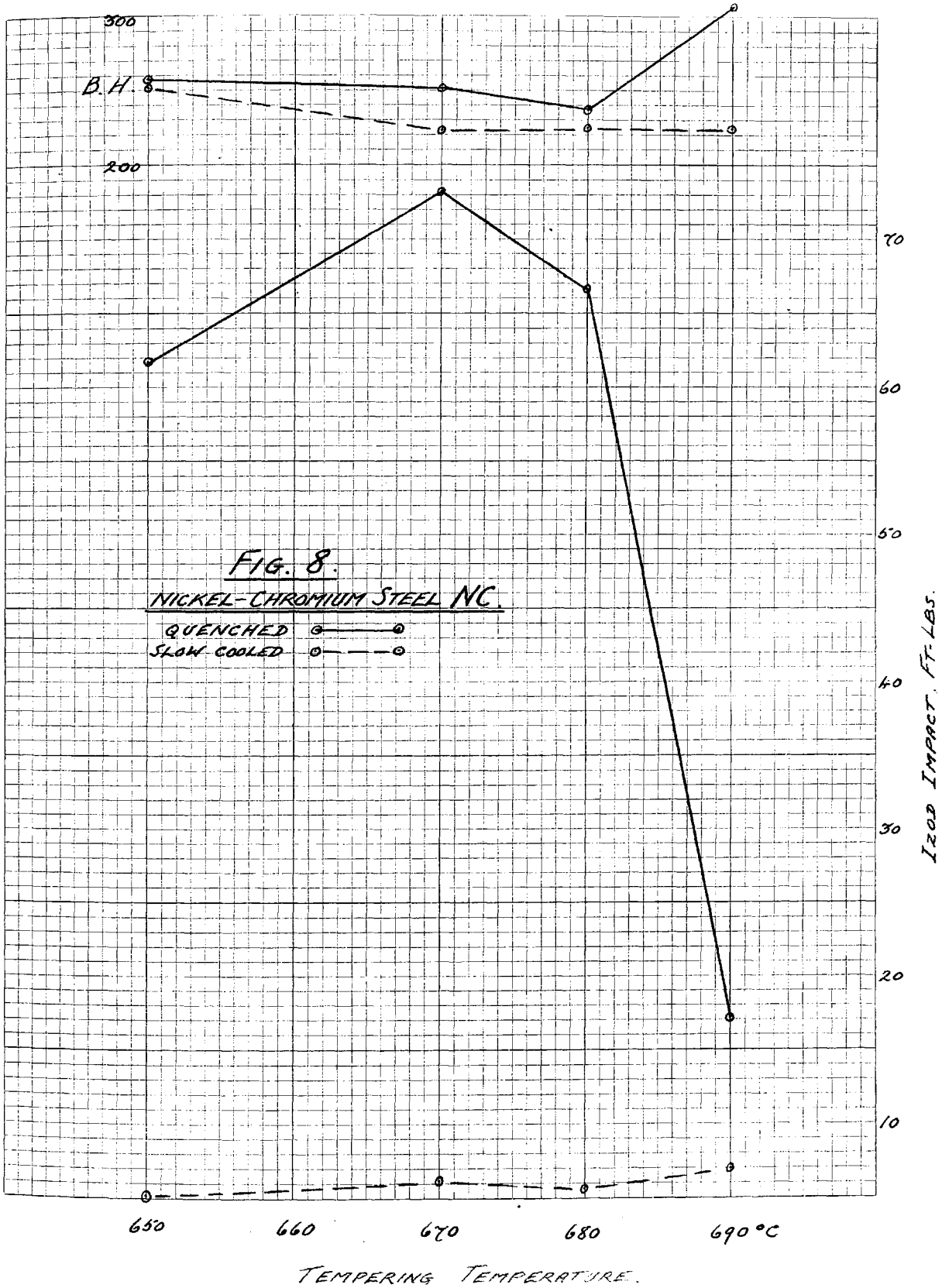


FIG. 7.
NICKEL-CHROMIUM STEEL AH.
B.H.
QUENCHED
SLOW COOLED
400
300
200
60
50
40
30
20
10
0
IZOD IMPACT, FT-LBS.
650
660
670
680
690°C.
TEMPERING TEMPERATURE.


300
B.H.N.
200
70
60
50
40
30
20
10
650
660
670
680
690 °C
TEMPERING TEMPERATURE.
IZOD IMPACT. FT-LBS.
FIG. 8.
NICKEL-CHROMIUM STEEL NC.
QUENCHED
SLOW COOLED

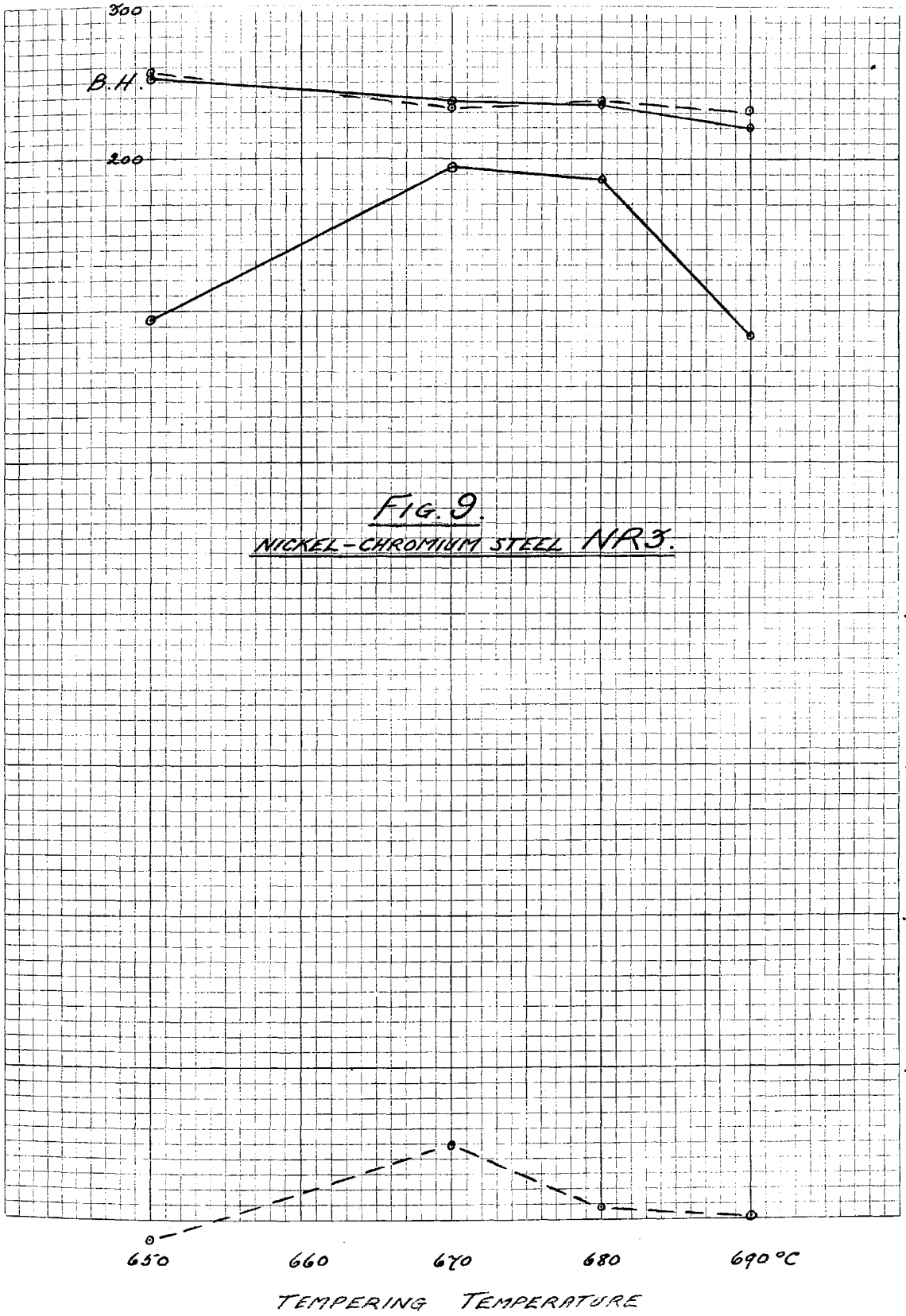
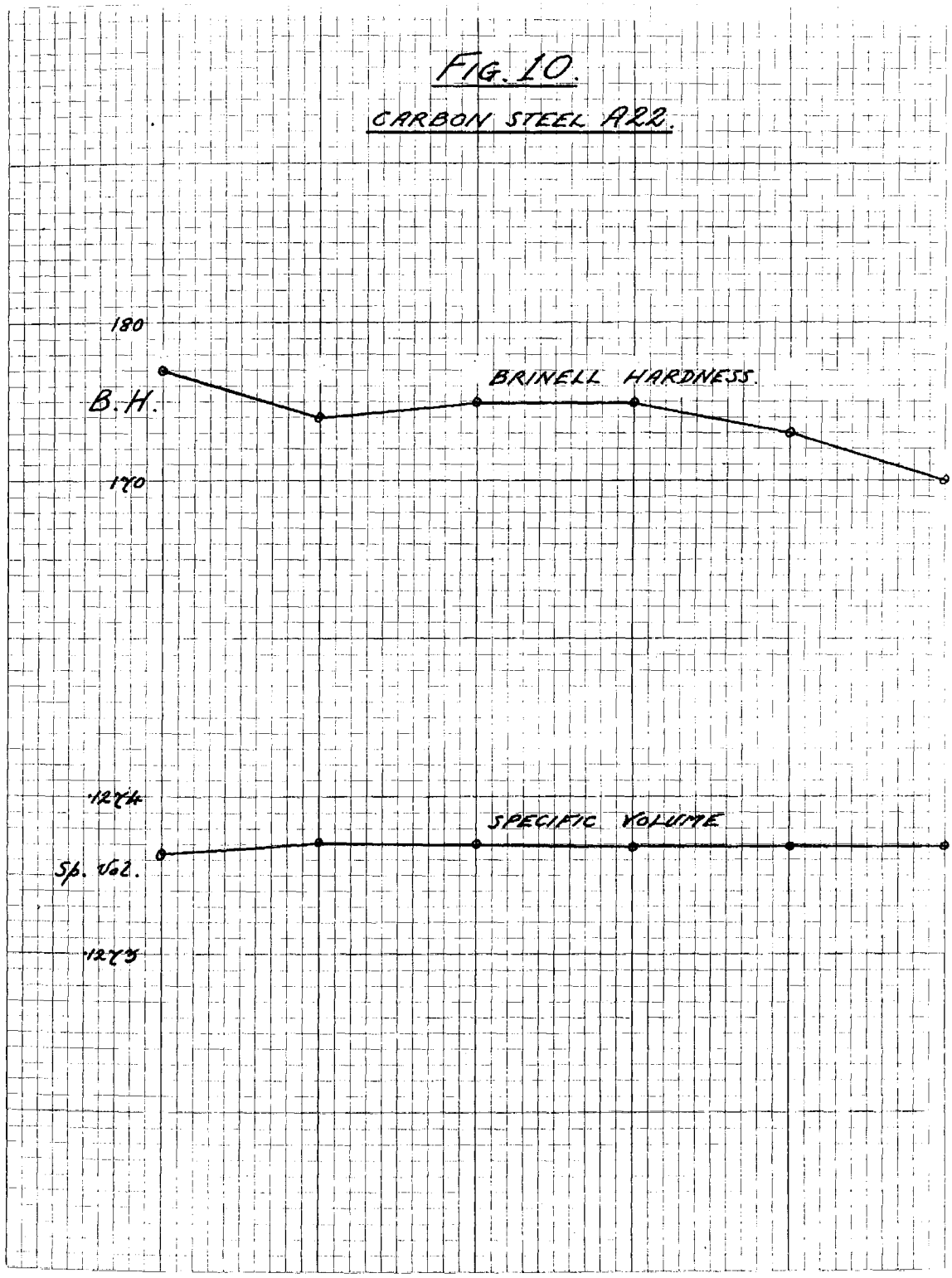


FIG. 9.

NICKEL-CHROMIUM STEEL NR3.

FIG. 10.

CARBON STEEL A22.



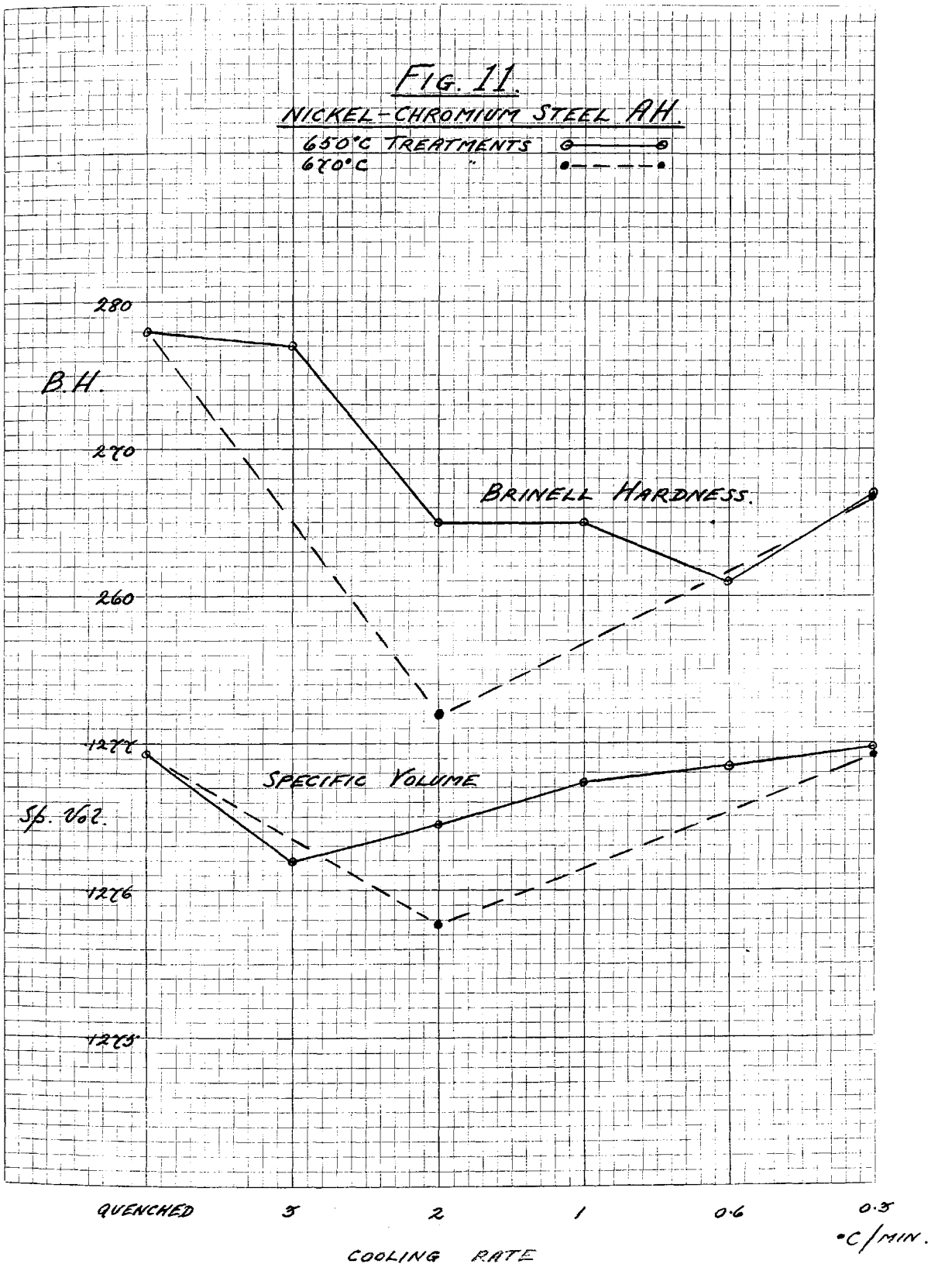


FIG. 11.
NICKEL-CHROMIUM STEEL AH.
650°C TREATMENTS
670°C
280
B.H.
270
260
BRINELL HARDNESS.
1277
Sp. Vol.
1276
1275
SPECIFIC VOLUME
QUENCHED
3
2
1
0.6
0.3
°C/MIN.
COOLING RATE

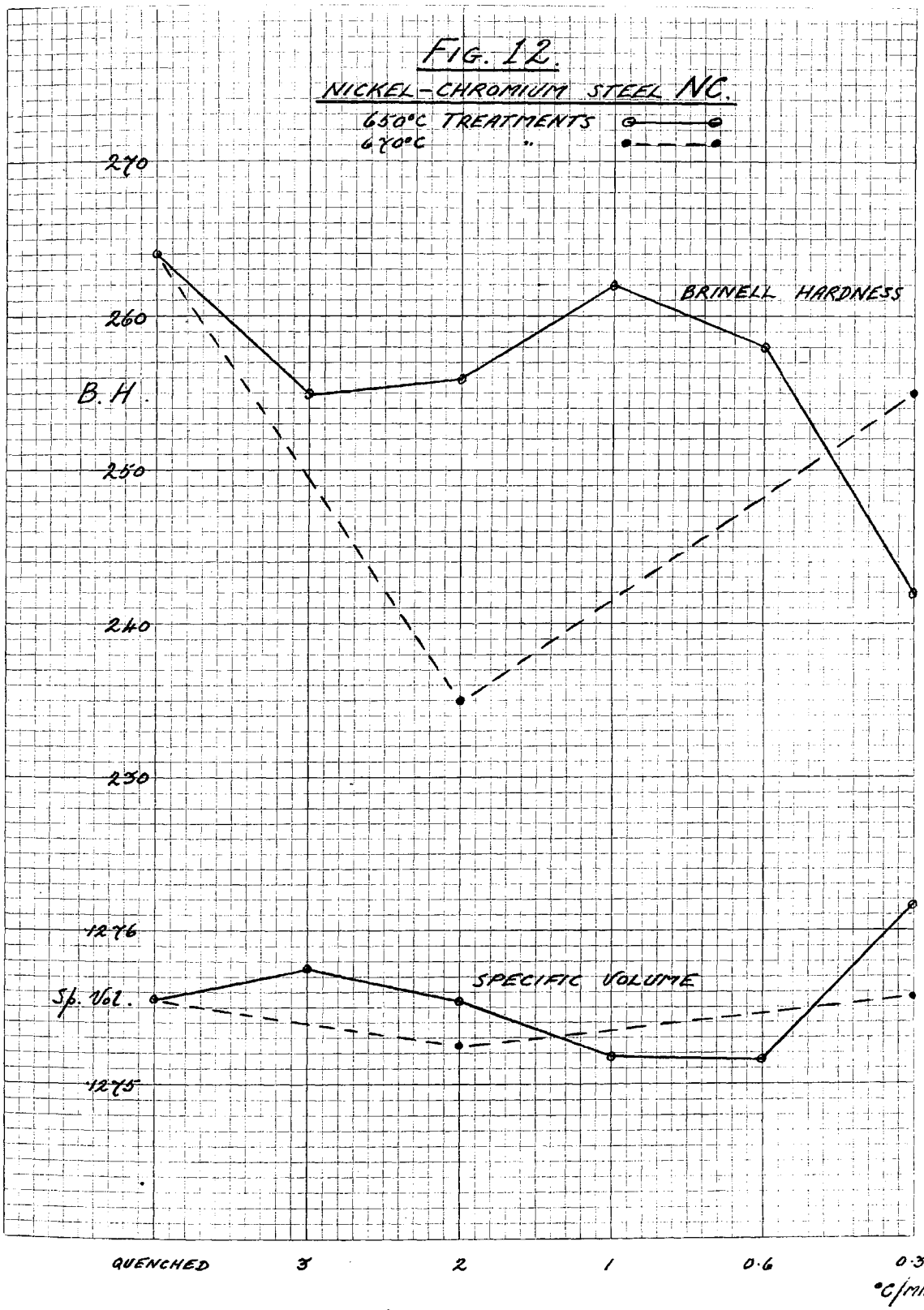


FIG. 12.
NICKEL-CHROMIUM STEEL NC.
650°C TREATMENTS
670°C "
270
260
B.H.
250
240
230
BRINELL HARDNESS
1·276
Sp. Vol.
1·275
SPECIFIC VOLUME
QUENCHED
3
2
1
0·6
0·3
°C/MIN.
COOLING RATE

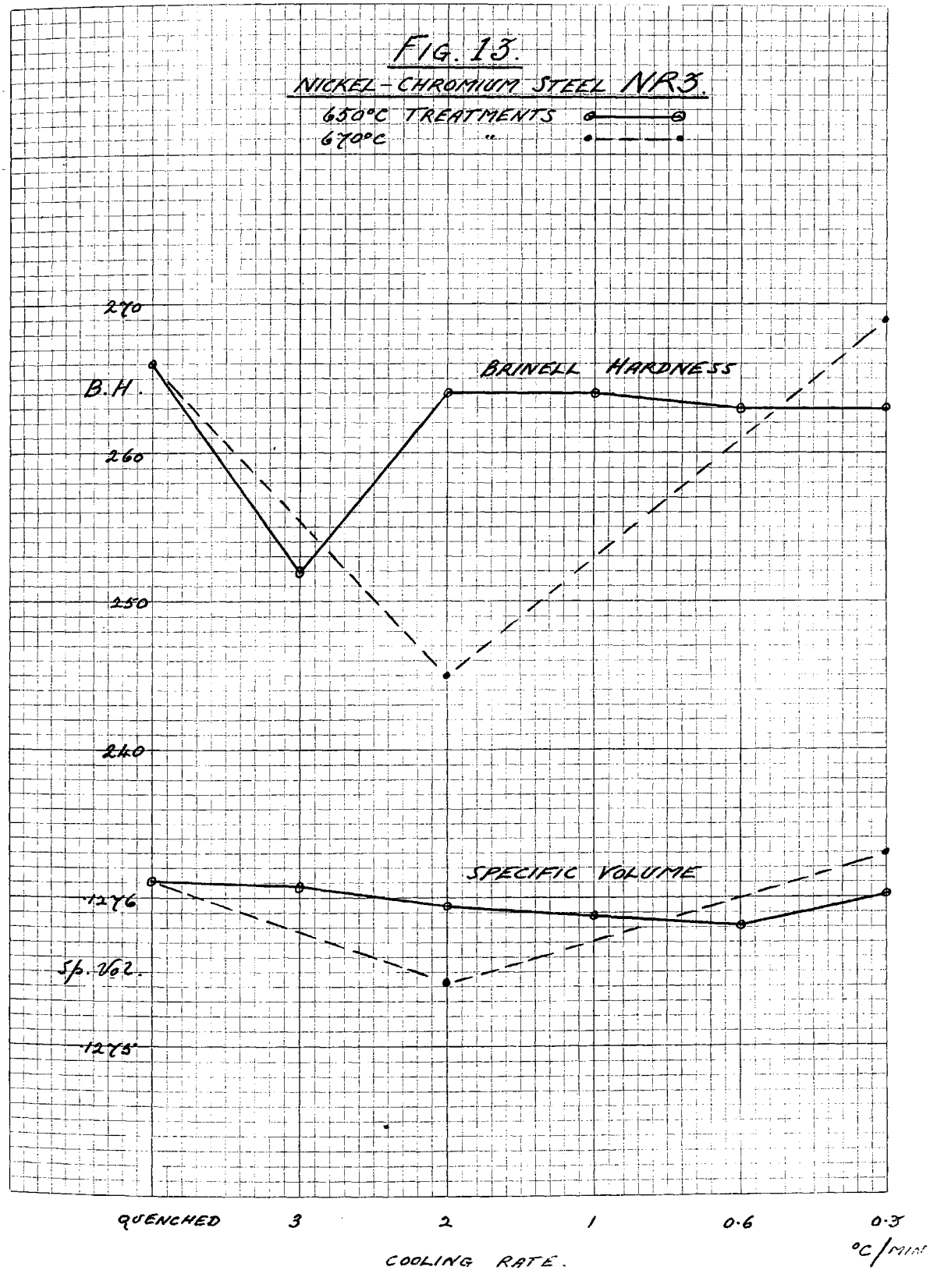


FIG. 13.
NICKEL-CHROMIUM STEEL NR3.
650°C TREATMENTS
670°C "
270
B.H.
260
250
240
BRINELL HARDNESS
SPECIFIC VOLUME
·1276
Sp. Vol.
·1275
QUENCHED
3
2
1
0·6
0·3
°C/MIN.
COOLING RATE.

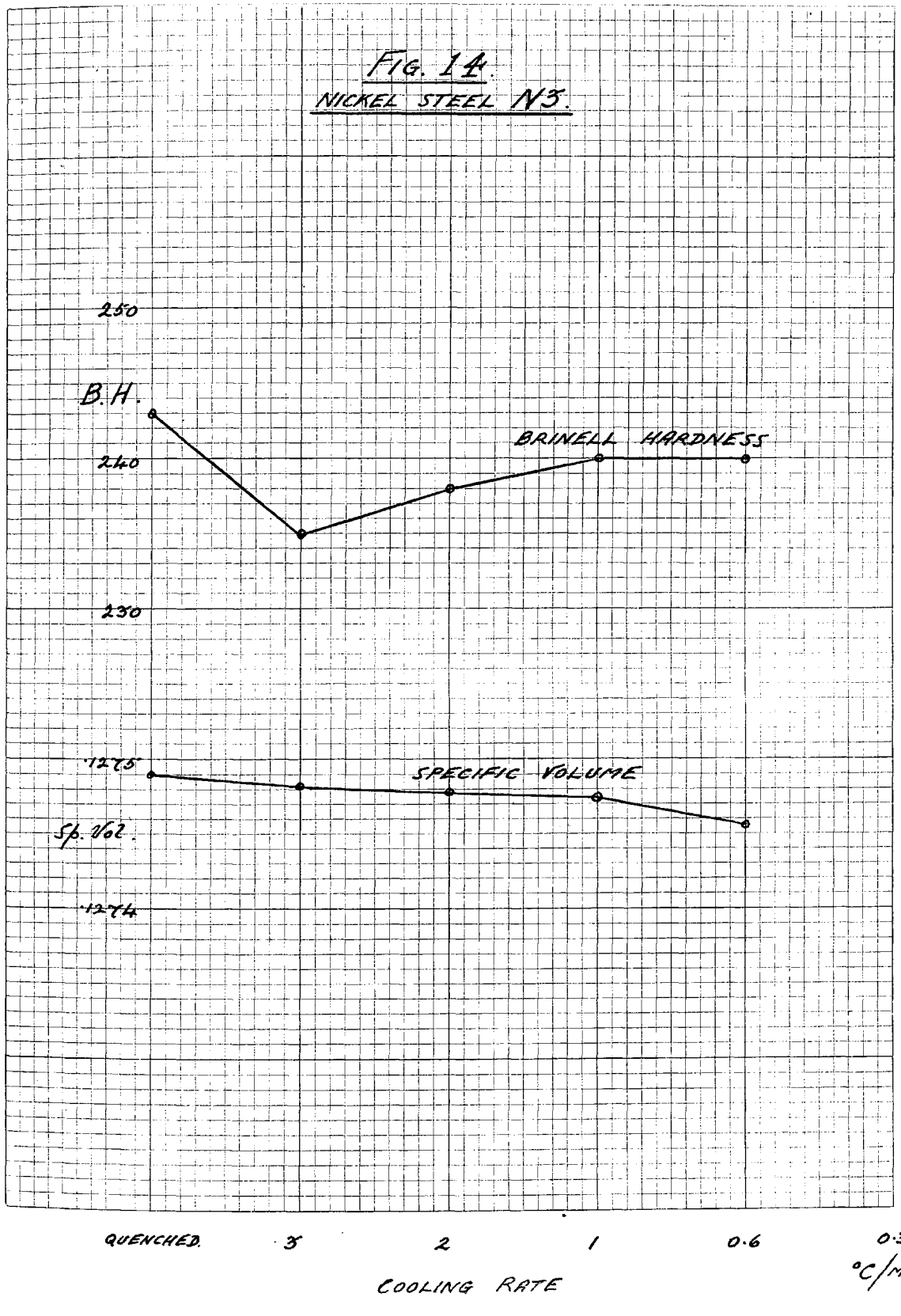
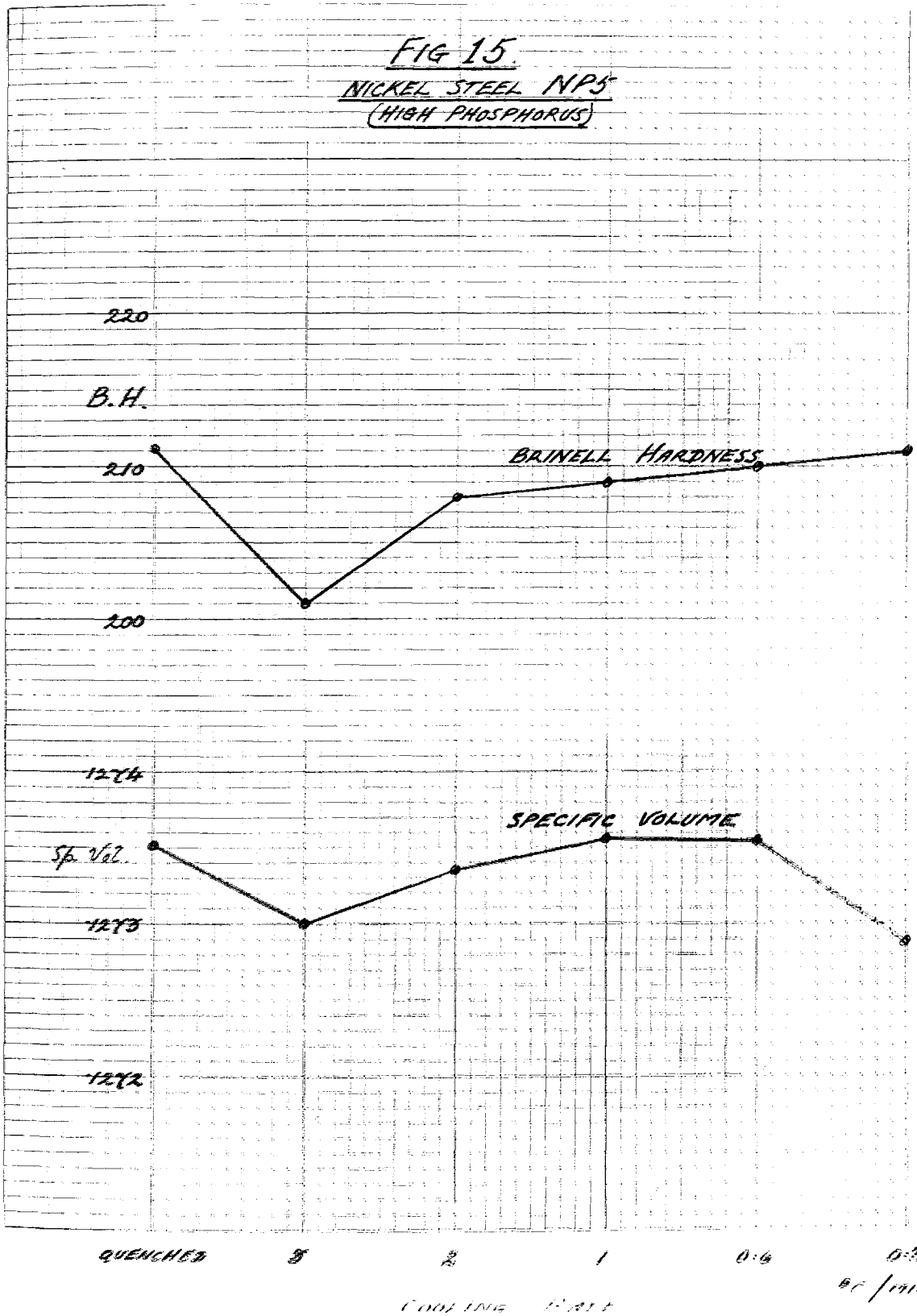


FIG. 14.
NICKEL STEEL N3.
250
B.H.
240
BRINELL HARDNESS
230
·1275
SPECIFIC VOLUME
Sp. Vol.
·1274
QUENCHED.
3
2
1
0·6
0·3
°C/MIN.
COOLING RATE

FIG 15.

NICKEL STEEL NP5
(HIGH PHOSPHORUS)

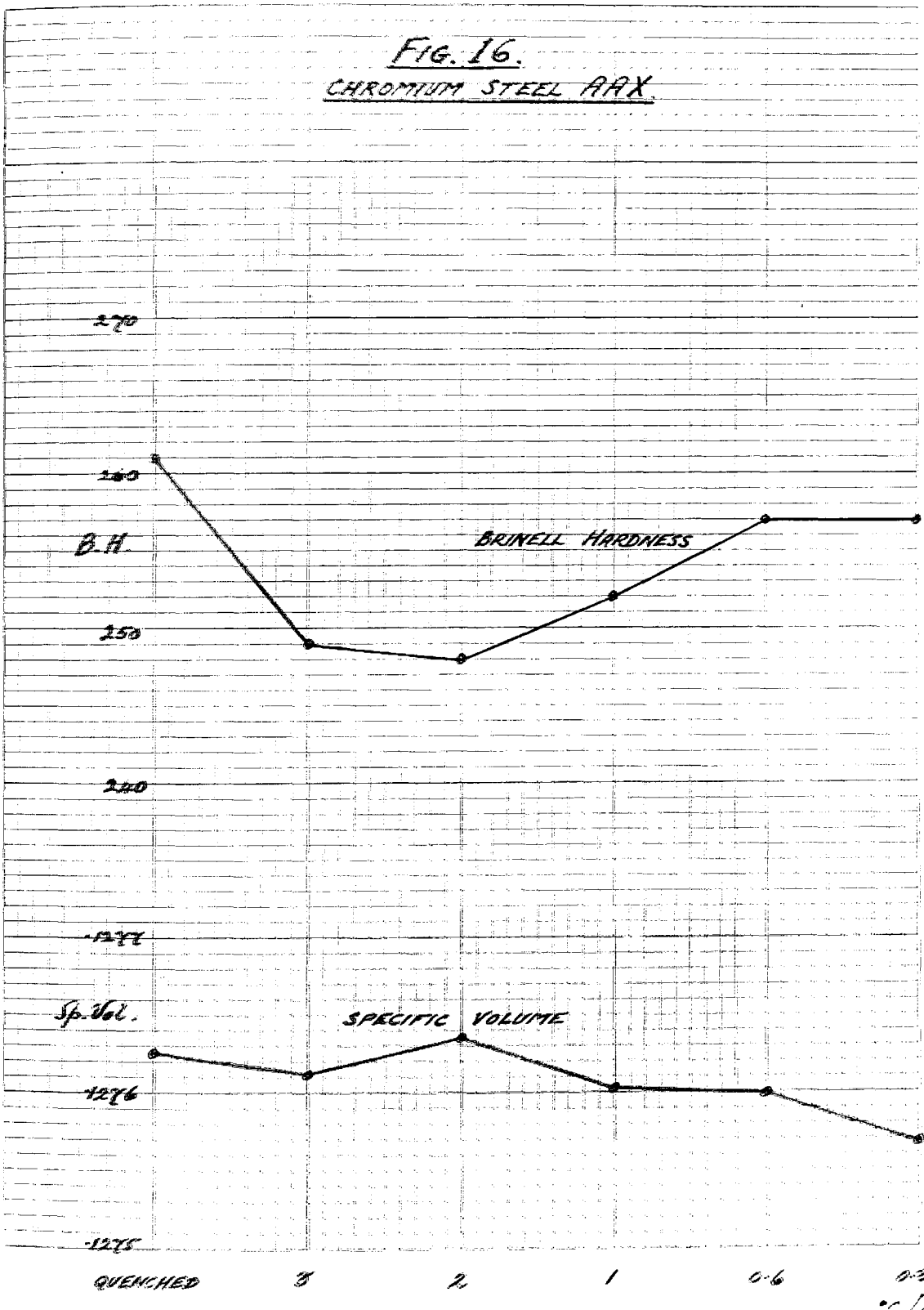


BF / 1911

COOLING RATE

FIG. 16.

CHROMIUM STEEL AAX.



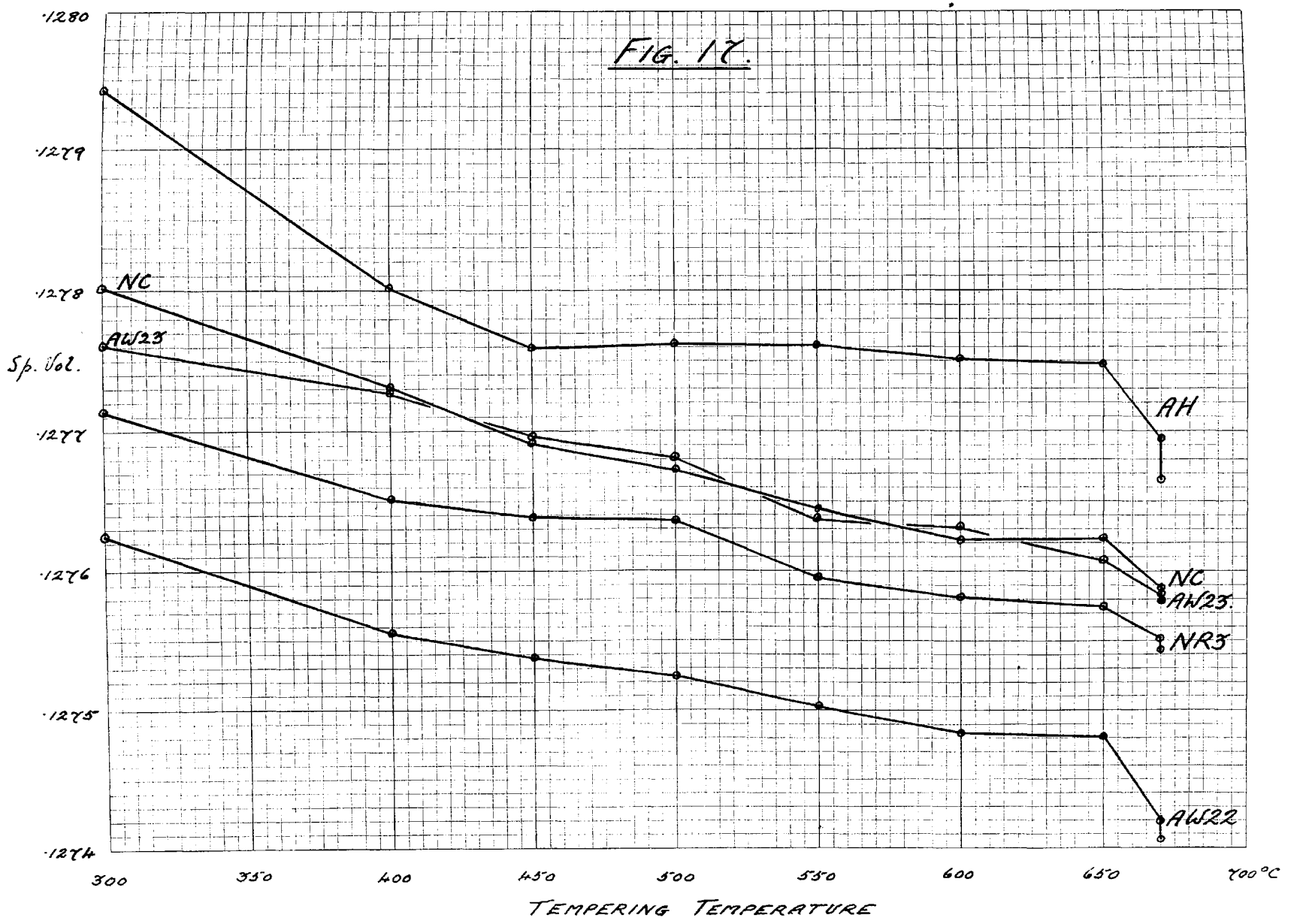


FIG. 17.
·1280
·1279
·1278
·1277
·1276
·1275
·1274
Sp. Vol.
NC
AW23
AH
NC
AW23.
NR3
AW22
300
350
400
450
500
550
600
650
700°C
TEMPERING TEMPERATURE

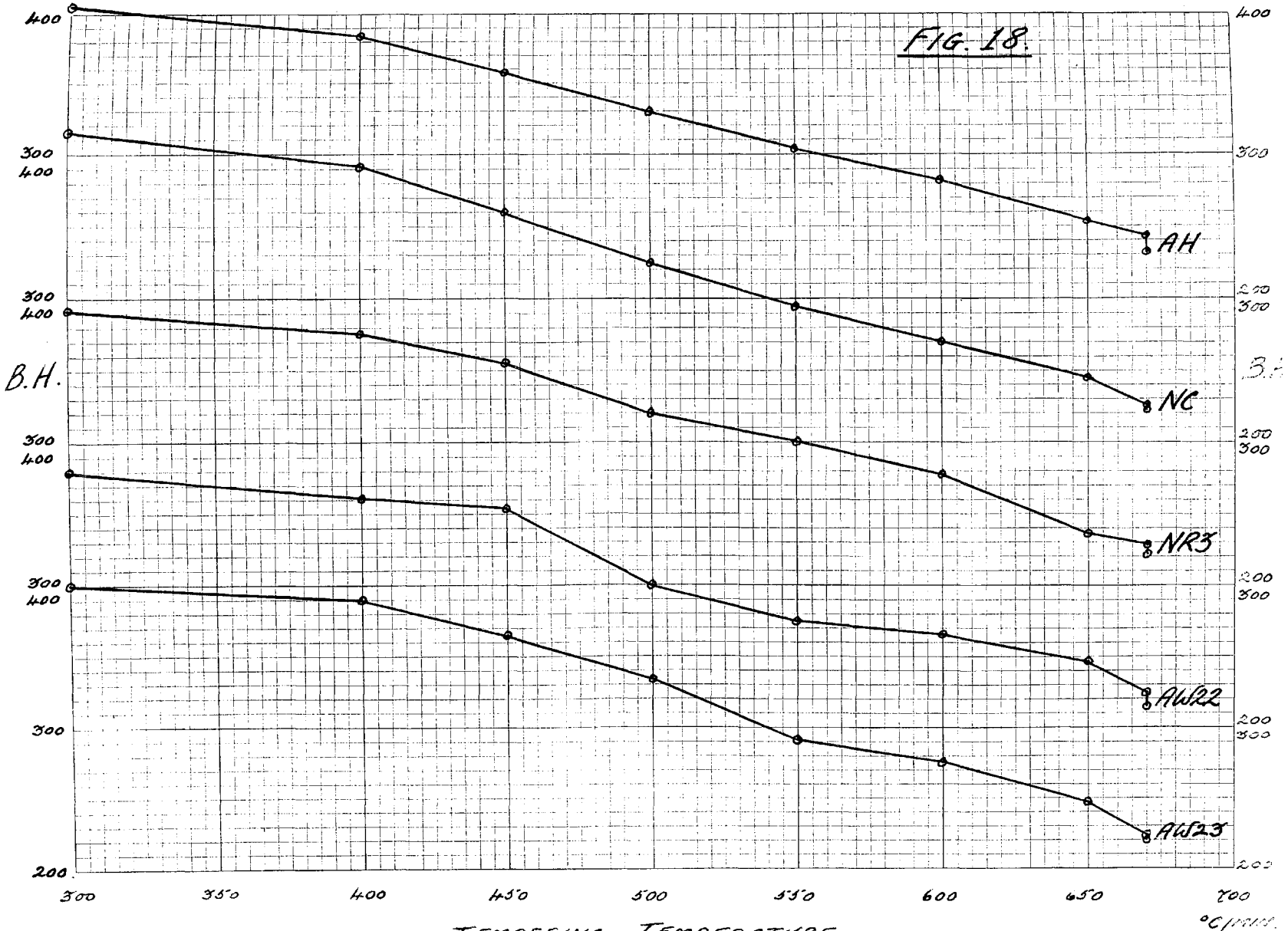
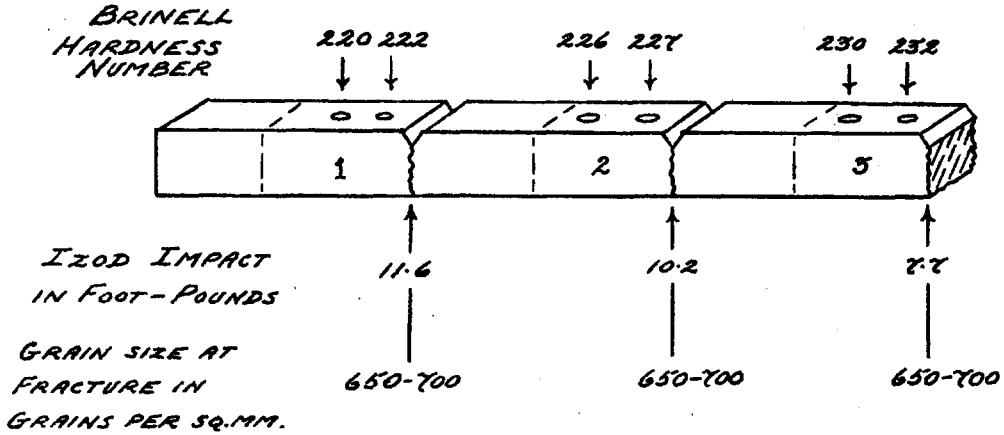


FIG. 18.
B.H.
AH
NC
NR3
AW22
AW23
400
300
200
300
350
400
450
500
550
600
650
700
TEMPERING TEMPERATURE
°C/min.

FIG. 19.



(The notches are drawn all on one face to avoid confusion.)

FIG. 20.

NICKEL STEEL N3. x370.
ETCHED IN BOILING SODIUM PICRATE.

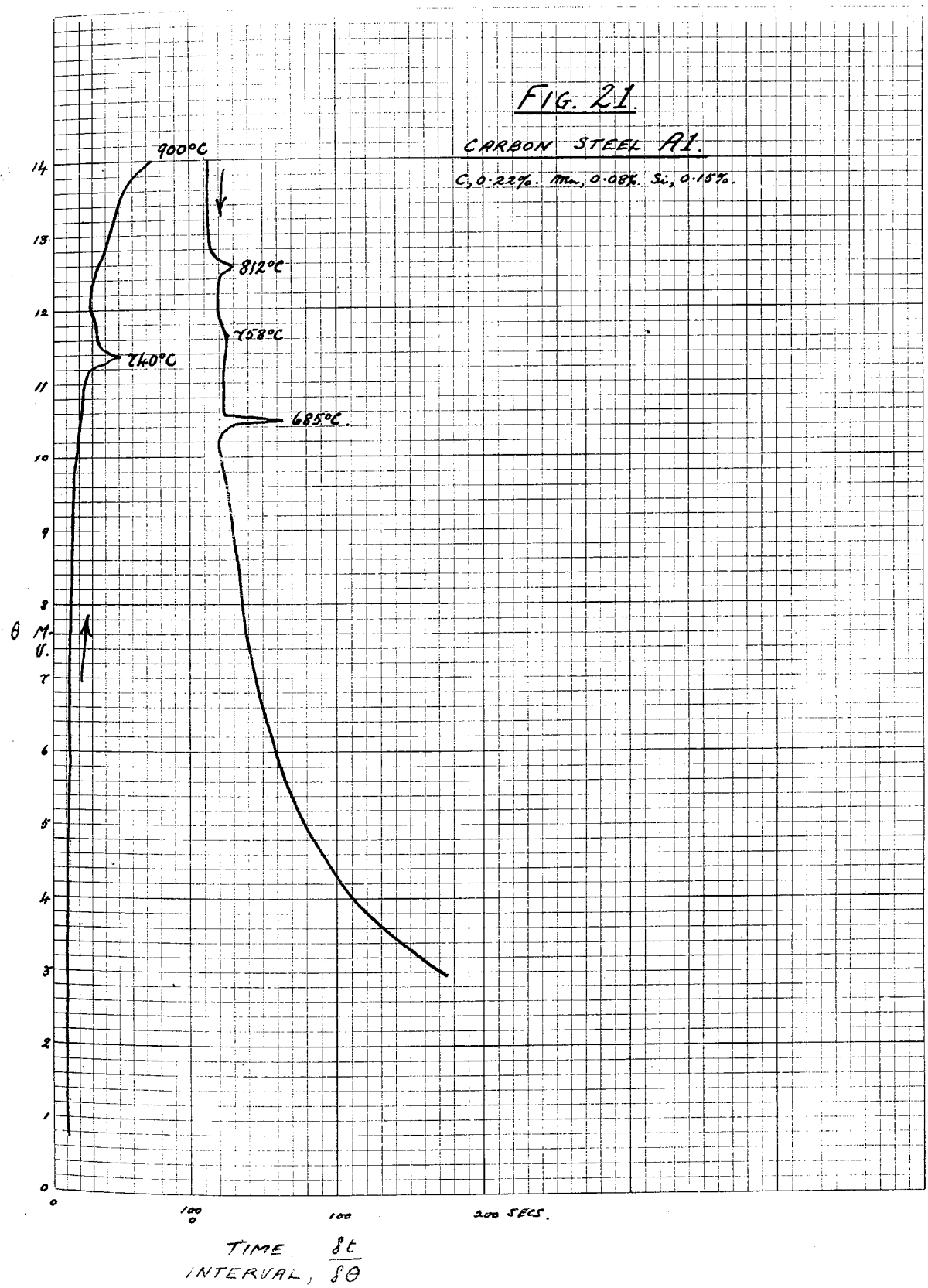


FIG. 21.
CARBON STEEL A1.
C, 0·22%. Mn, 0·08%. Si, 0·15%.
900°C
812°C
758°C
740°C
685°C.
θ M.V.
0
1
2
3
4
5
6
7
8
9
10
11
12
13
14
100
0
100
200 SECS.
TIME INTERVAL, δt/δθ

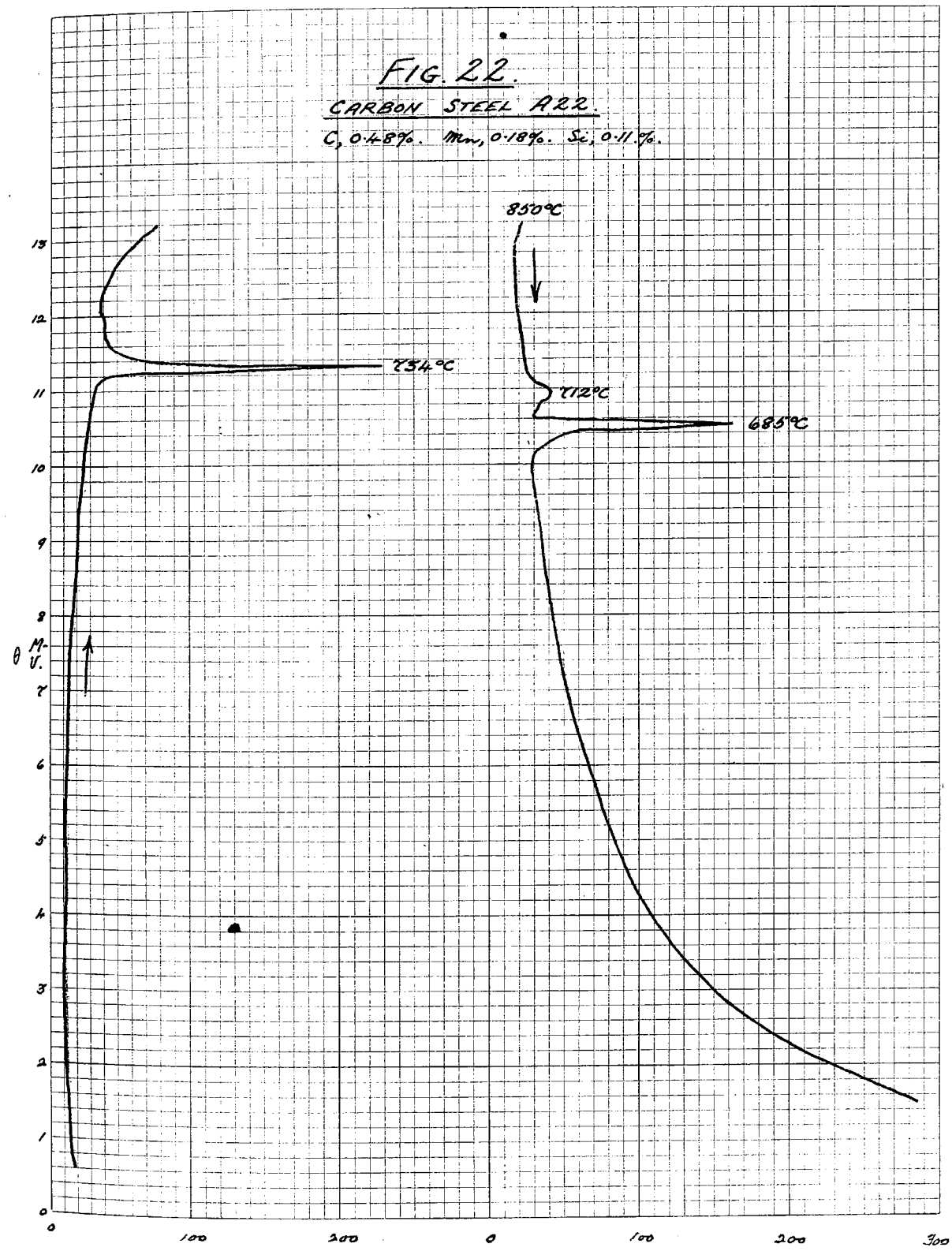
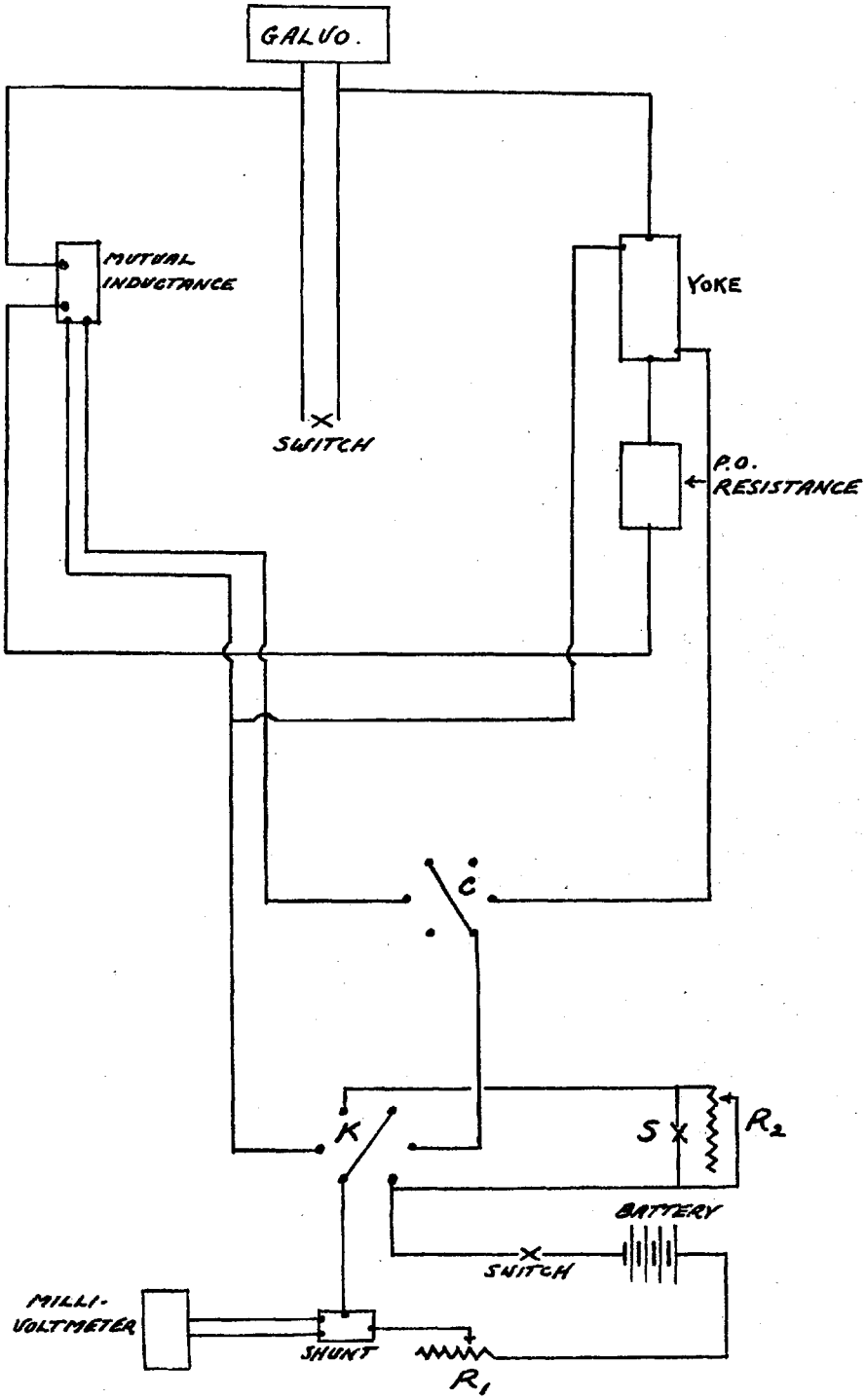


FIG. 22.
CARBON STEEL A22.
C, 0·48%. Mn, 0·18%. Si, 0·11%.
850°C
734°C
712°C
685°C
θ M-V.
0
1
2
3
4
5
6
7
8
9
10
11
12
13
0
100
200
0
100
200
300 SECS.
TIME. 5t
INTERVAL, 5θ

FIG. 23.

APPARATUS FOR MAGNETIC TESTS.



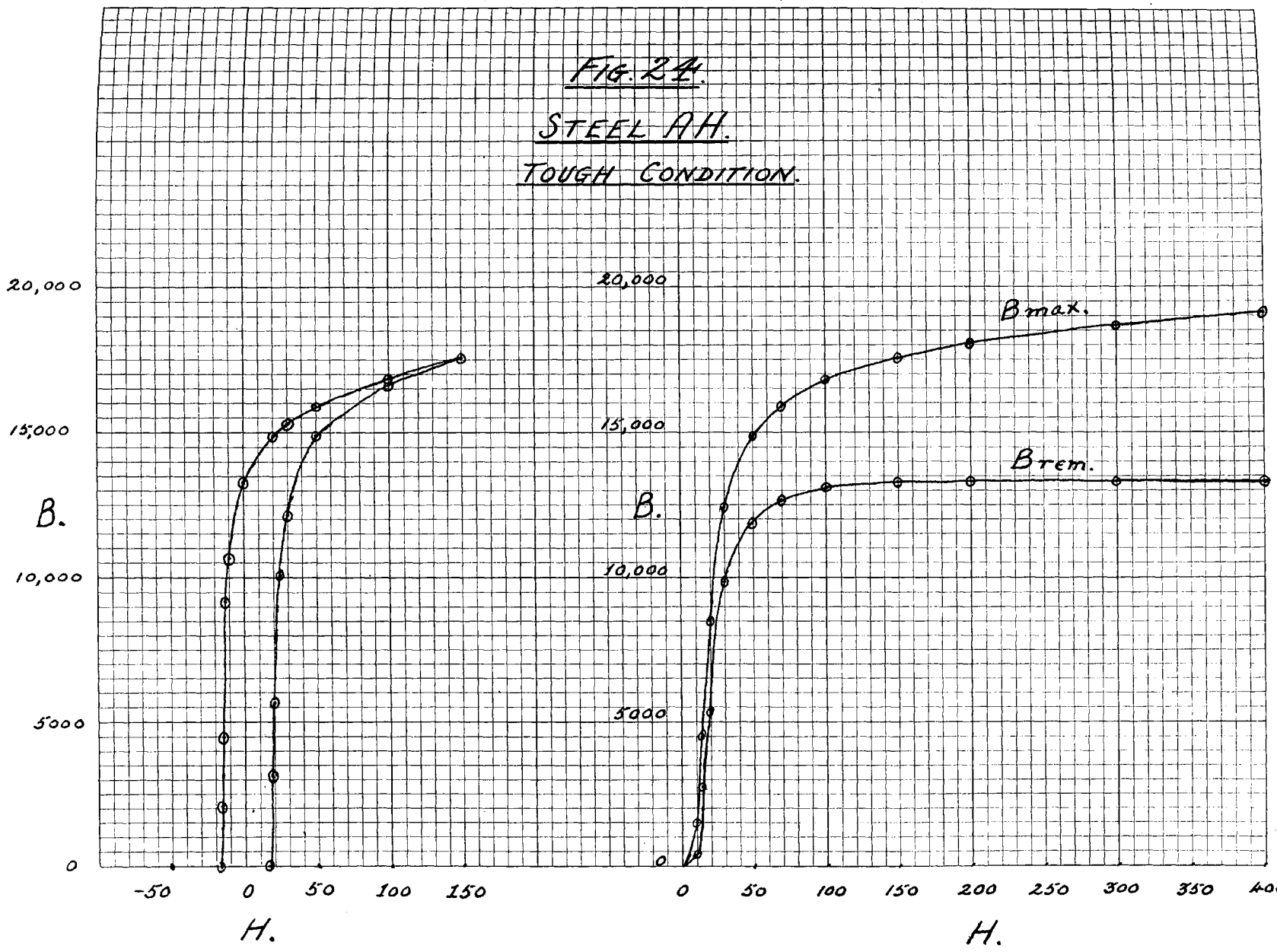


FIG. 24.
STEEL AH.
TOUGH CONDITION.
20,000
15,000
B.
10,000
5000
0
-50
0
50
100
150
H.
20,000
15,000
B.
10,000
5000
0
Bmax.
Brem.
0
50
100
150
200
250
300
350
400
H.

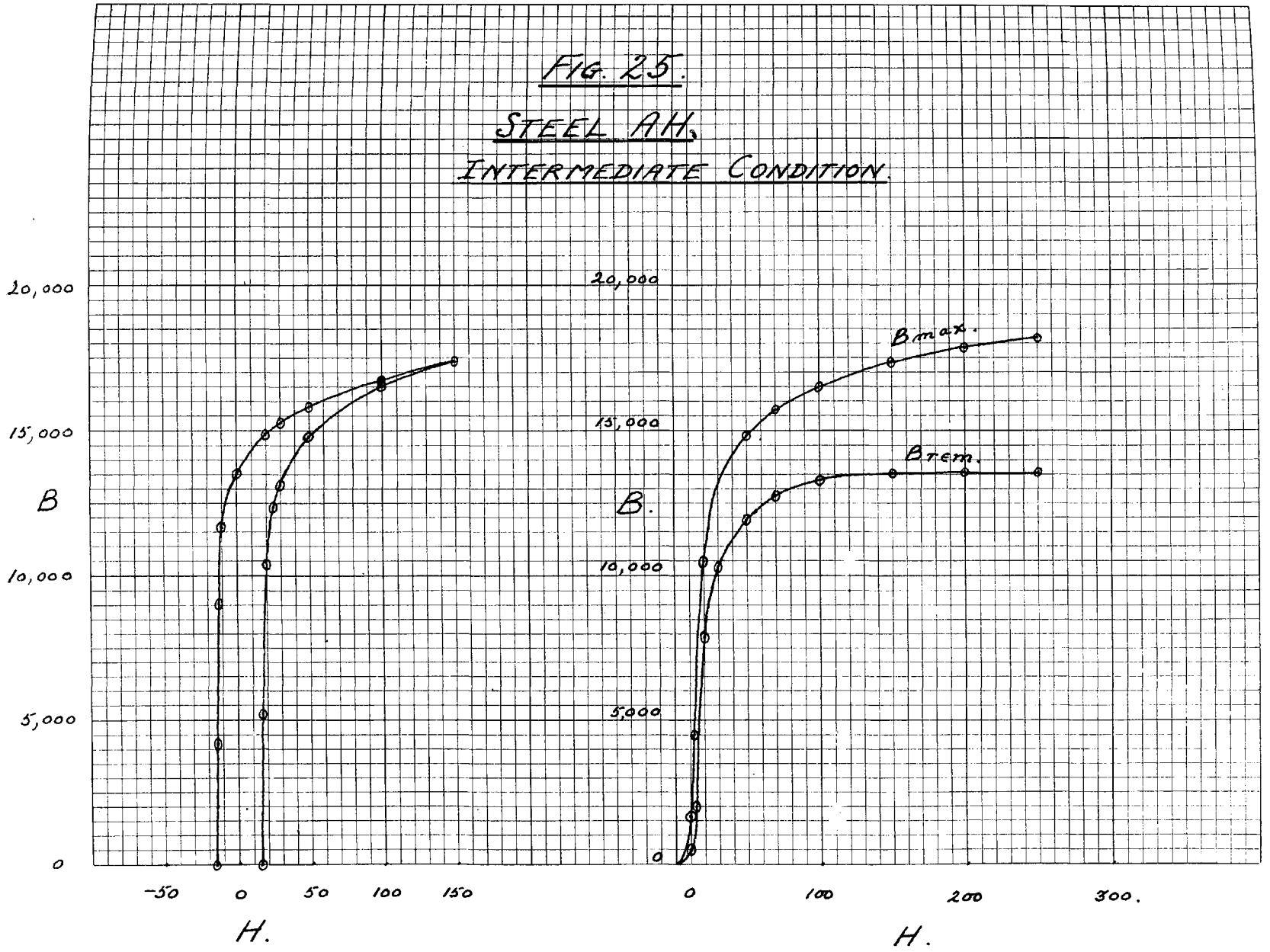


FIG. 25.
STEEL AH.
INTERMEDIATE CONDITION.
20,000
15,000
B
10,000
5,000
0
-50
0
50
100
150
H.
20,000
15,000
B.
10,000
5,000
0
B max.
B rem.
0
100
200
300.
H.

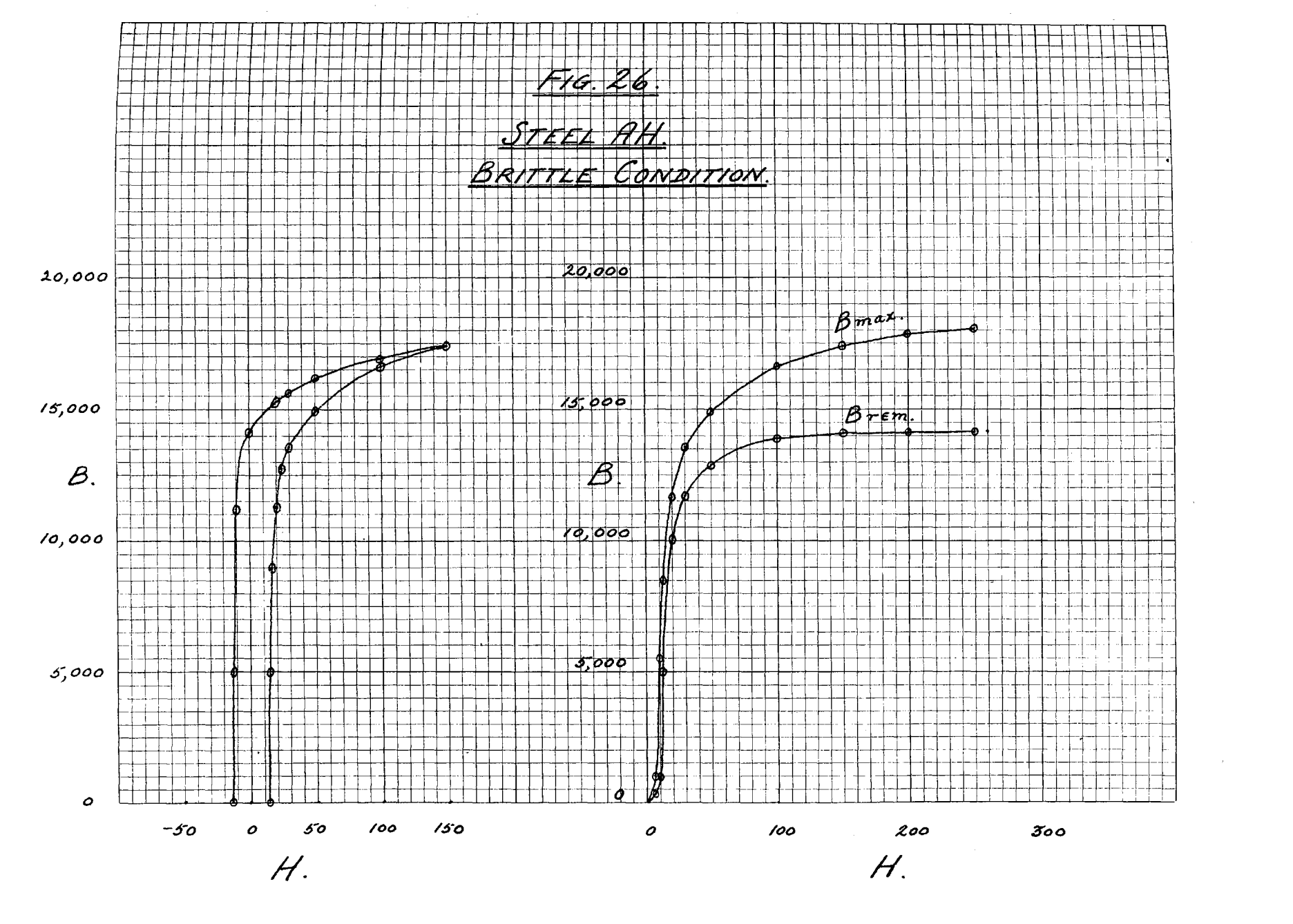
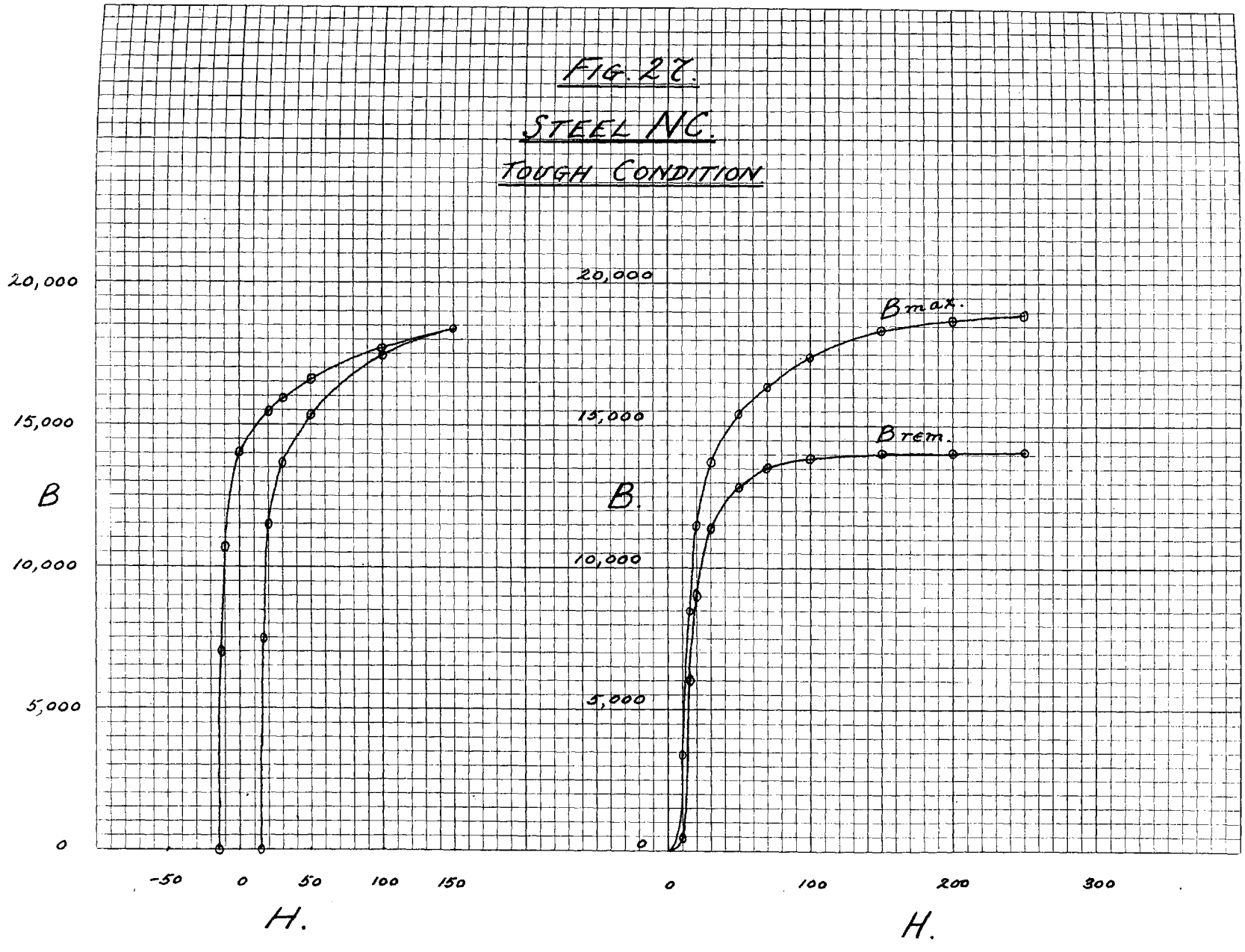


FIG. 26.
STEEL AH.
BRITTLE CONDITION.
20,000
15,000
B.
10,000
5,000
0
-50
0
50
100
150
H.
20,000
15,000
B.
10,000
5,000
0
Bmax.
Brem.
0
100
200
300
H.

FIG. 27.

STEEL NC.

TOUGH CONDITION



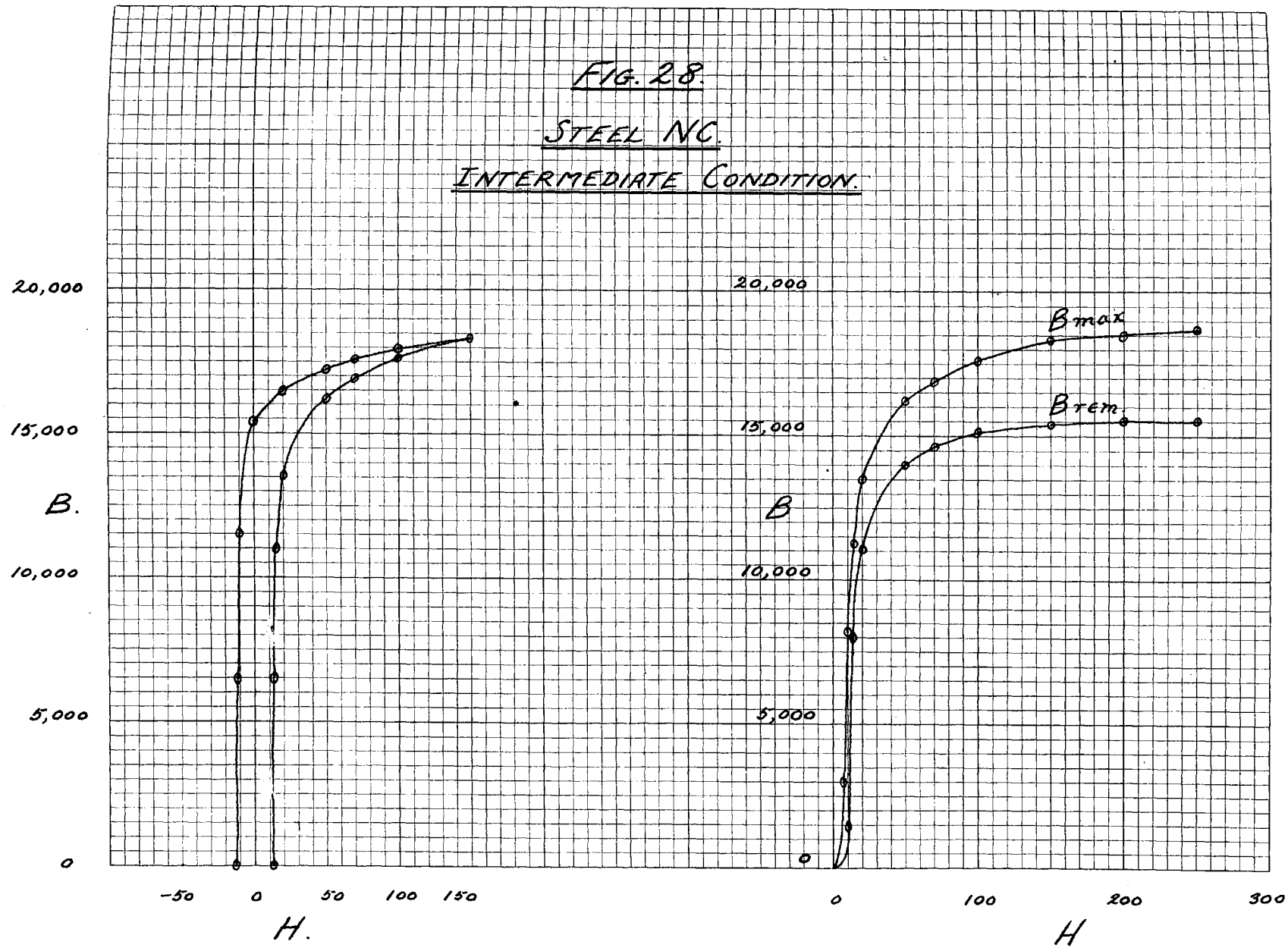


FIG. 28.
STEEL NC.
INTERMEDIATE CONDITION.
20,000
15,000
B.
10,000
5,000
0
-50
0
50
100
150
H.
20,000
15,000
B
10,000
5,000
0
0
100
200
300
H
Bmax
Brem.

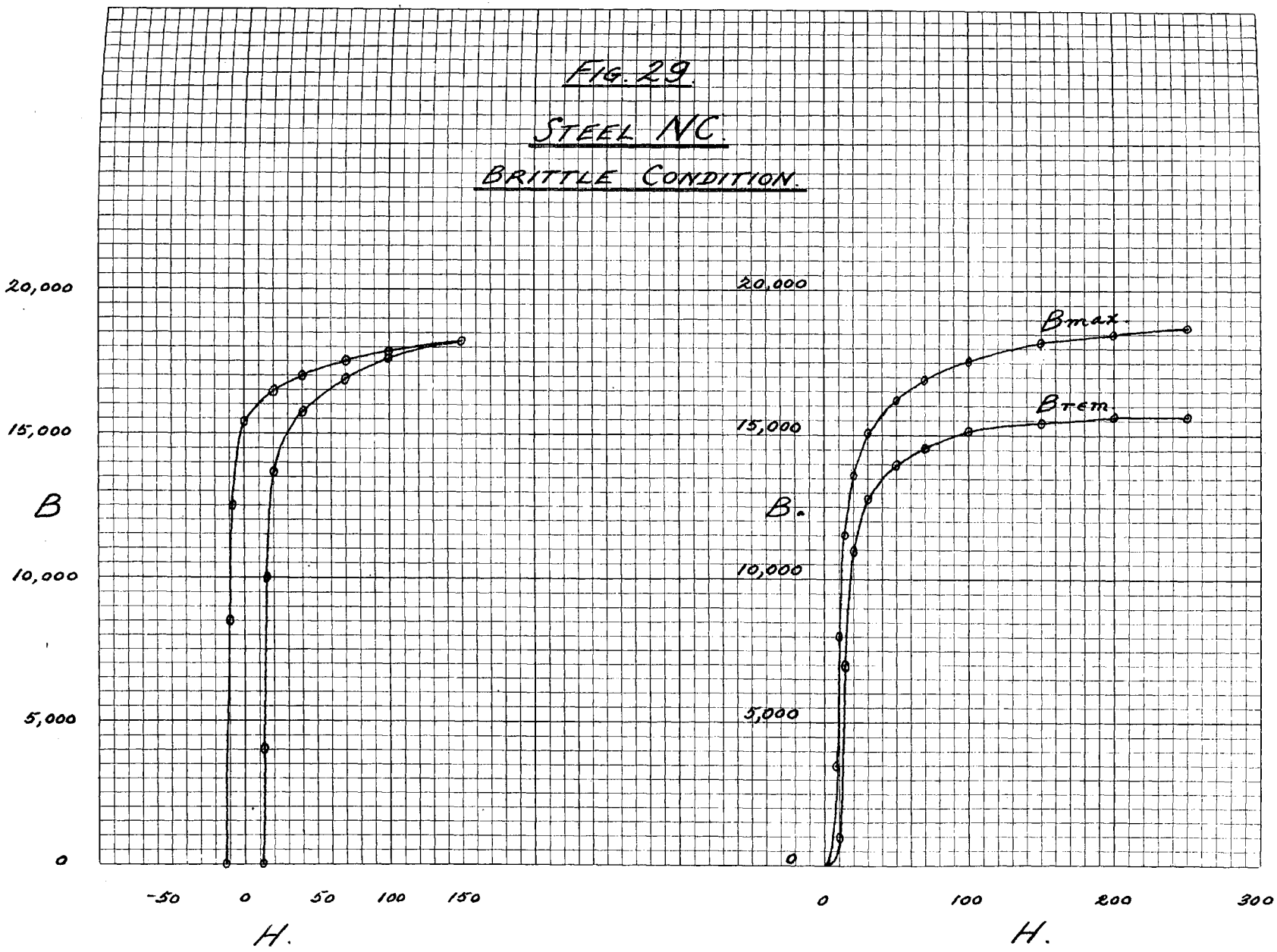


FIG. 29
STEEL NC.
BRITTLE CONDITION.
20,000
15,000
B
10,000
5,000
0
-50
0
50
100
150
H.
20,000
15,000
B.
10,000
5,000
0
Bmax.
Brem.
0
100
200
300
H.

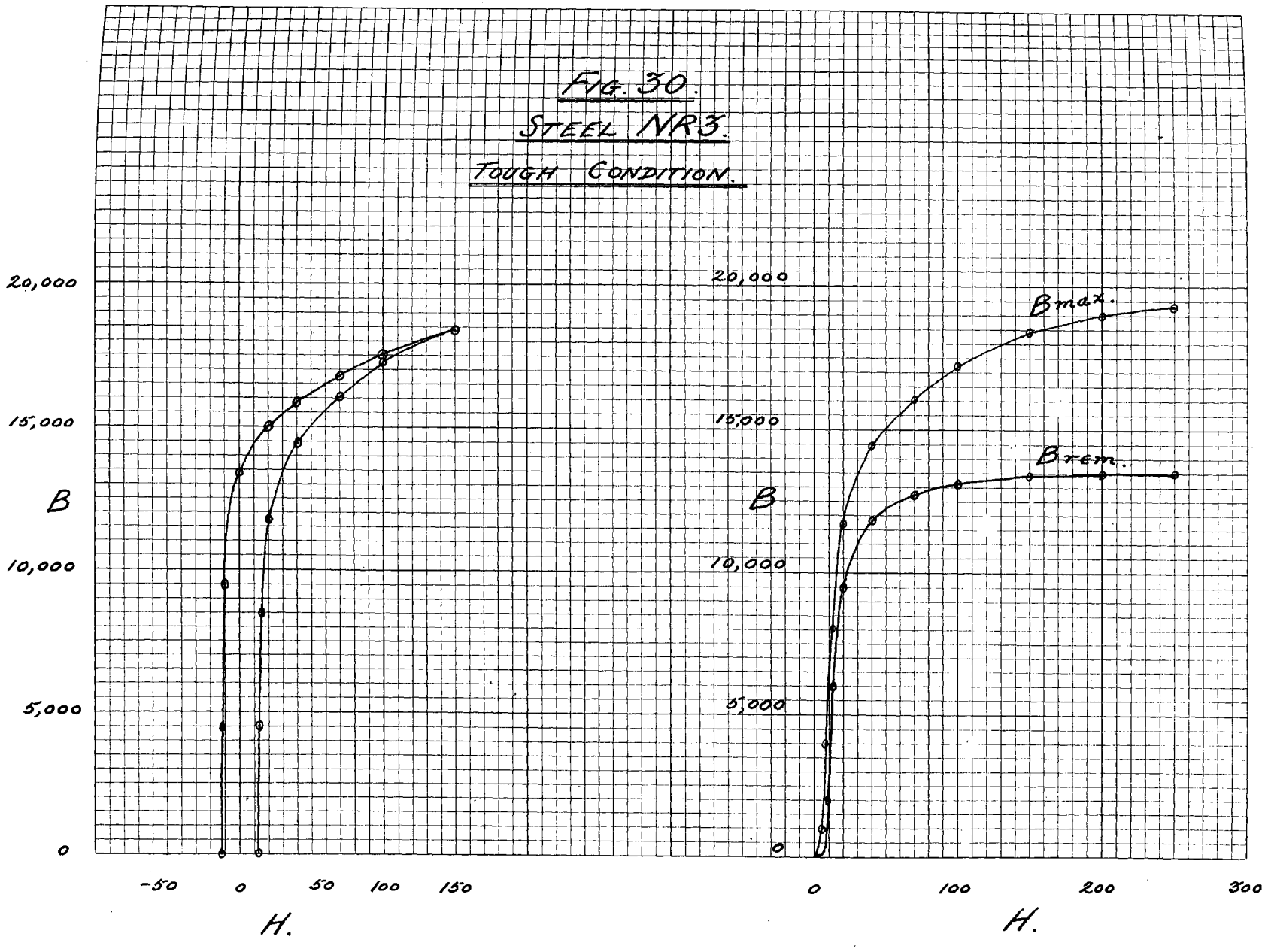
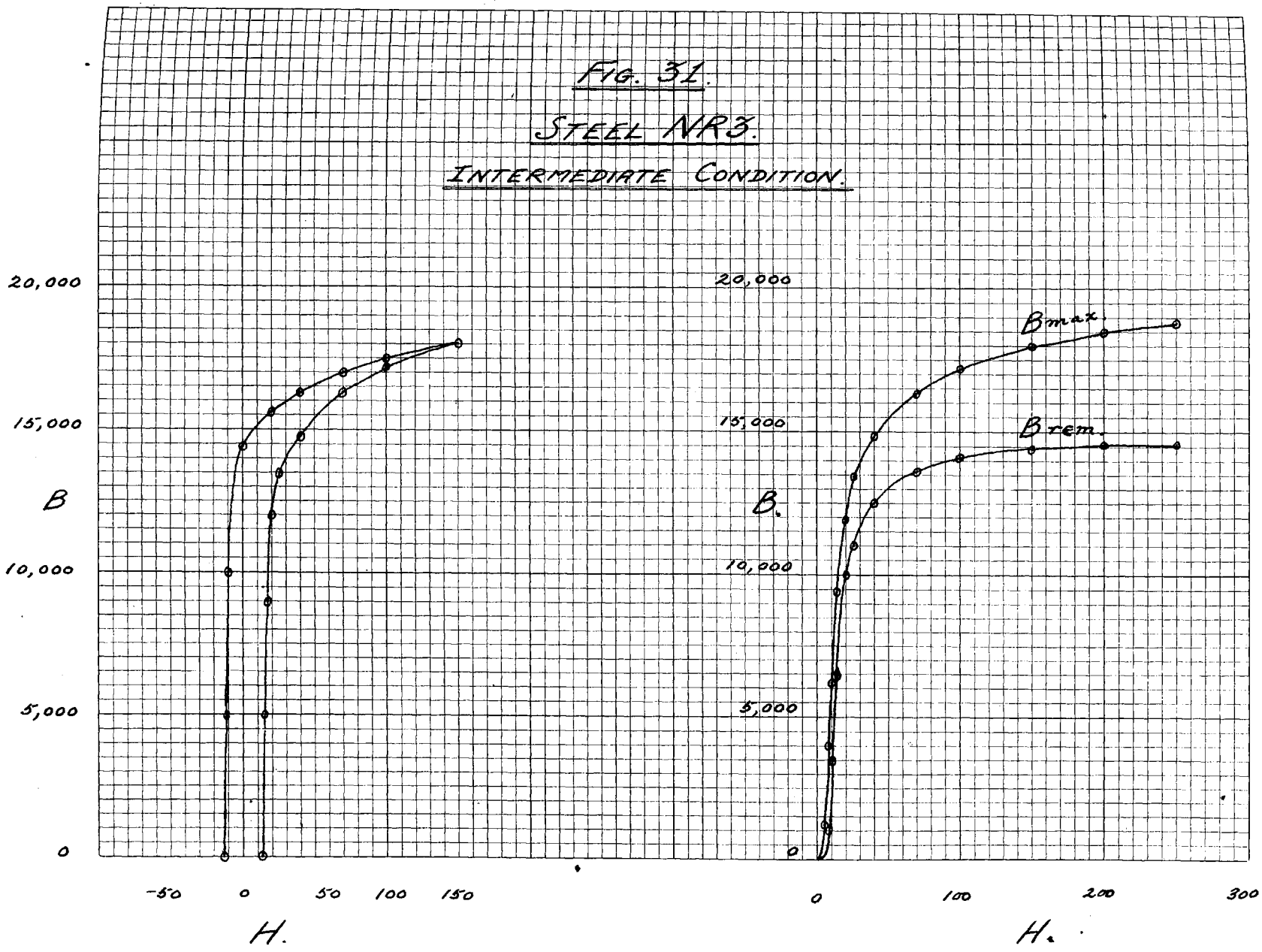


FIG. 30.
STEEL NR3.
TOUGH CONDITION.
20,000
15,000
B
10,000
5,000
0
-50
0
50
100
150
H.
20,000
15,000
B
10,000
5,000
0
Bmax.
Brem.
0
100
200
300
H.

FIG. 31.

STEEL NR3.

INTERMEDIATE CONDITION.



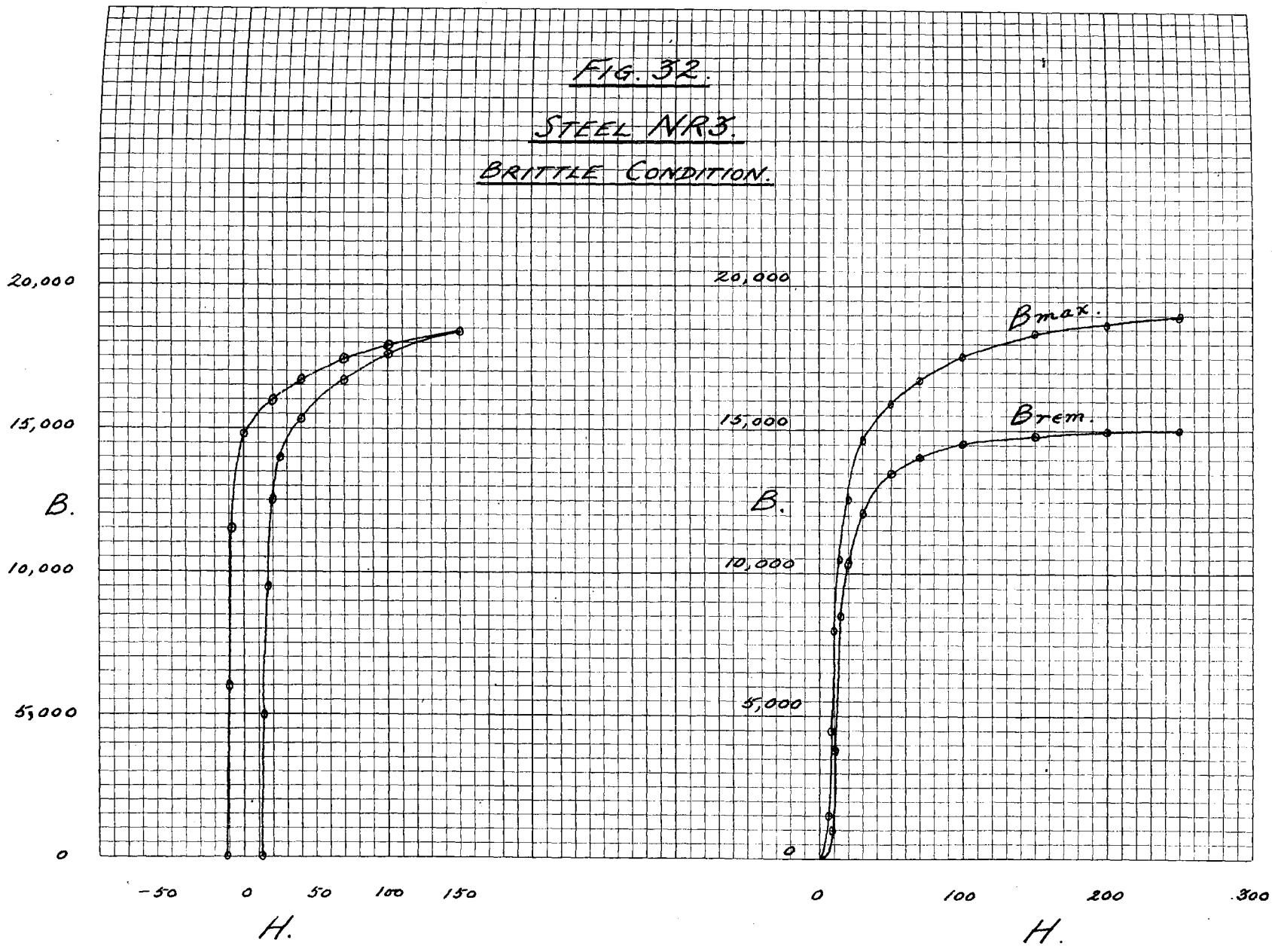


FIG. 32.
STEEL NR3.
BRITTLE CONDITION.
20,000
15,000
B.
10,000
5,000
0
-50
0
50
100
150
H.
20,000
15,000
B.
10,000
5,000
0
Bmax.
Brem.
0
100
200
300
H.

PART II

TABLE I.

Steel	12 Hours at 540°C Sp. Vol.	24 Hours at 540°C. Sp. Vol.	B.H.	12 Hours at 590°C Sp. Vol.	B.H.	12 Hours at 640°C Cp. Vol.	B.H.
A4.	0·127516	0·127518	239	0·127521	222	0·127518	212
2.	0·127331	0·127357	188	0·127377	156	0·127369	145
4.	0·127441	0·127477	234	0·127481	201	0·127491	182
6.	0·127641	0·127696	266	0·127706	209	0·127703	222

Steel	12 Hours at 660°C Sp.Vol.	B.H.	12 Hours at 675°C Sp.Vol.	B.H.	12 Hours at 685°C Sp.Vol.	B.H.	12 Hours at 695°C Sp.Vol.	B.H.
A4.	0·127519	201	0·127514	191	0·127509	184	0·127488	166
2.	0·127361	156	0·127355	143	0·127438	222	0·127548	342
4.	0·127481	170	0·127428	170	0·127440	197	0·127630	355
6.	0·127700	209	0·127644	194	0·127755	300	0·128423	652

Steel	12 Hours at 700°C Sp.Vol.	B.H.	12 Hours at 705°C Sp.Vol.	B.H.	24 Hours at 705°C Sp.Vol.	B.H.	12 Hours at 710°C Sp.Vol.	B.H.
A4.	0·127514	170	0·127517	143	0·127507	143	0·127512	155
2.	0·127441	275	0·127340	243	0·127318	185	0·127179	176
4.	0·127405	222	0·127329	201	0·127295	175	0·127249	170
6.	0·128107	326	0·128713	239	0·128918	209	0·129304	182

Steel	12 Hours at 715°C Sp.Vol.	B.H.
A4.	0·127532	153
2.	0·127565	240
4.	0·127275	172
6.	0·129874	209

TABLE 2.

Steel	6 Hours at 670°C Sp. Vol.	6 Hours at 675°C Sp. Vol.	6 Hours at 680°C Sp. Vol.	6 Hours at 685°C Sp. Vol.	6 Hours at 690°C Sp. Vol.
A4.	0·127530	0·127530	0·127523	0·127488	0·127521
2.	0·127213	0·127207	0·127212	0·127242	0·127440
4.	0·127443	0·127381	0·127362	0·127519	0·127709
6.	0·127741	0·127698	0·127700	0·128096	0·128207

Steel	6 Hours at 695°C Sp. Vol.	6 Hours at 700°C Sp. Vol.	6 Hours at 705°C Sp. Vol.	3 Hours at 710°C Sp. Vol.	3 Hours at 715°C Sp. Vol.
A4.	0·127529	0·127500	0·127535	0·127502	0·127514
2.	0·127506	0·127504	0·127364	0·127402	0·127418
4.	0·127842	0·127680	0·127380	0·127285	0·127271
6.	0·128560	0·128923	0·128899	0·128534	0·128941

Steel	3 Hours at 720°C Sp. Vol.	3 Hours at 725°C Sp. Vol.	3 Hours at 730°C Sp. Vol.	3 Hours at 735°C Sp. Vol.	3 Hours at 740°C Sp. Vol.
A4.	0·127513	0·127569	0·127506	0·127644	0·127540
2.	0·127429	0·127432	0·127466	0·127450	0·127451
4.	0·127685	0·127777	0·127906	0·127962	0·128162
6.	0·129509	0·129548	0·129647	0·129666	0·129665.

TABLE 3.

Steel	6 hours at 665°C. Sp.Vol.	B.H.	6 hours at 670°C. Sp. Vol.	B.H.	6 hours at 675°C. Sp. Vol.	B.H.	6 hours at 680°C. Sp. Vol.	B.H.
NP1	0·127106	195	0·127092	192	0·127037	192	0·127008	194
NP3	0·127229	207	0·127164	203	0·127106	201	0·127061	200
NP5	0·127278	217	0·127221	211	0·127131	212	0·127082	211
AW22	0·127395	218	0·127398	220	0·127400	221	0·127372	228
AW23	0·127585	221	0·127568	212	0·127558	205	0·127536	205

Steel	6 hours at 685°C. Sp.Vol.	B.H.	6 hours at 690°C. Sp.Vol.	B.H.	6 hours at 695°C. Sp.Vol.	B.H.	6 hours at 700°C. Sp. Vol.	B.H.
NP1	0·127001	202	0·127018	231	0·127094	243	0·127036	203
NP3	0·127067	223	0·127118	249	0·126997	228	0·127203	288
NP5	0·127022	210	0·127077	222	0·127121	253	0·126998	203
AW22	0·127351	232	0·127350	230	0·127438	238	0·127464	252
AW23	0·127517	204	0·127530	208	0·127554	213	0·127572	222

Steel	6 hours at 705°C. Sp.Vol.	B.H.	6 hours at 710°C. Sp.Vol.	B.H.	6 hours at 715°C. Sp.Vol.	B.H.	6 hours at 720°C. Sp.Vol.	B.H.
NP1	0·127131	231	0·127265	244	0·127300	278	0·127318	318
NP3	0·127322	298	0·127452	339	0·127524	353	0·127544	363
NP5	0·127000	218	0·127141	264	0·127298	302	0·127442	345
AW22	0·127483	258	0·127518	271	0·127551	288	0·127528	278
AW23	0·127577	223	0·127577	230	0·127578	241	0·127518	222

Steel	6 Hours at 725°C. Sp.Vol.	B.H.
NP1	0·127427	328
NP3	0·127614	397
NP5	0·127581	252
AW22	0·127608	361
AW23	0·127594	245

TABLE 4

Mechanical Tests on Nickel Steel, N.3

Specimen.	Treatment.	R.A.%	Elongation % on 2"	Maximum Strength Tons/sq.in.	Izod Impact Foot Pounds	Brinell Hardness.
1.	O.H. 850°C.,then tempered one week at 680°C-690°C and water quenched.	18·5	17·5	51·2	22·7	214
2.	Same as 1, but slow cooled at 0·3°C. per minute	17·9	23·4	46·4	18·3	193
3.	O.H.850°C., then tempered 4 days at 695°C-700°C and water quenched	6·45	8·8	78·4	8·3	237
4.	Same as 3, but slow cooled at 0·3°C per minute	9·0	11·0	56·4	9·8	227
5.	O.H. 820°C. and W.T. 620°C.	-	20-25	55·0	60	245

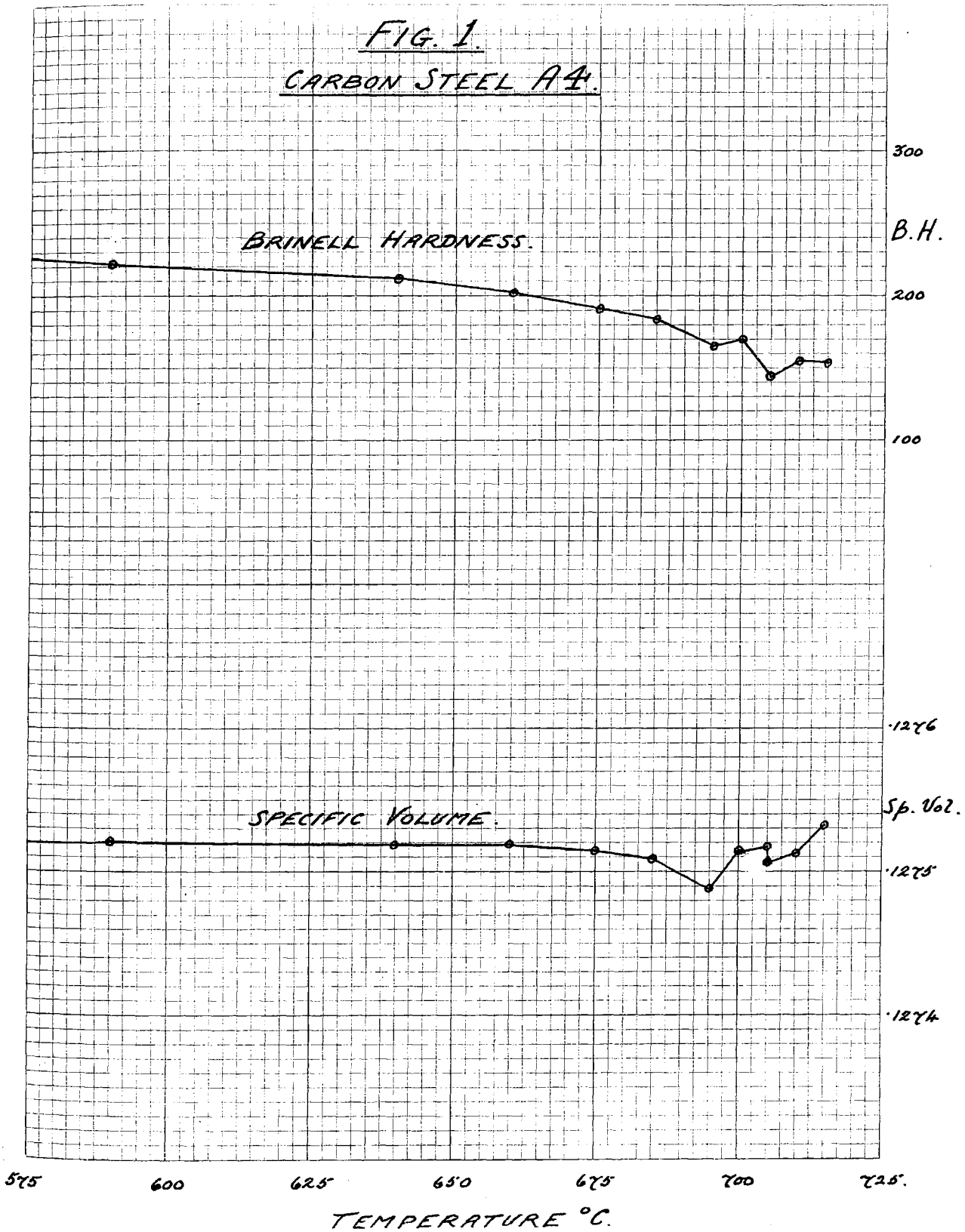


FIG. 1.
CARBON STEEL A4.
BRINELL HARDNESS.
300
B.H.
200
100
SPECIFIC VOLUME.
·1276
Sp. Vol.
·1275
·1274
575
600
625
650
675
700
725.
TEMPERATURE °C.

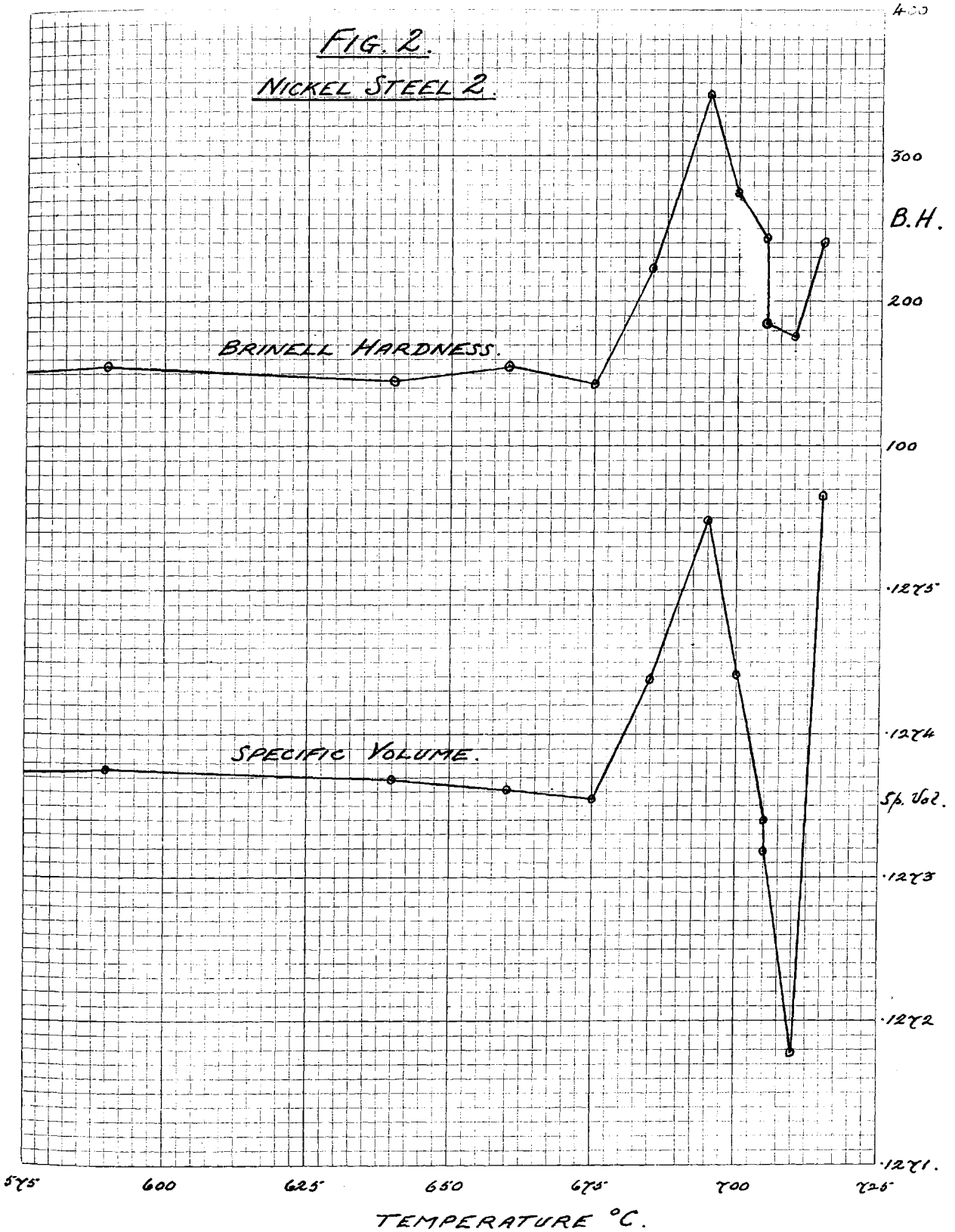


FIG. 2.
NICKEL STEEL 2.
BRINELL HARDNESS.
SPECIFIC VOLUME.
400
300
200
100
B.H.
·1275
·1274
·1273
·1272
·1271.
Sp. Vol.
575°
600
625°
650
675°
700
725°
TEMPERATURE °C.

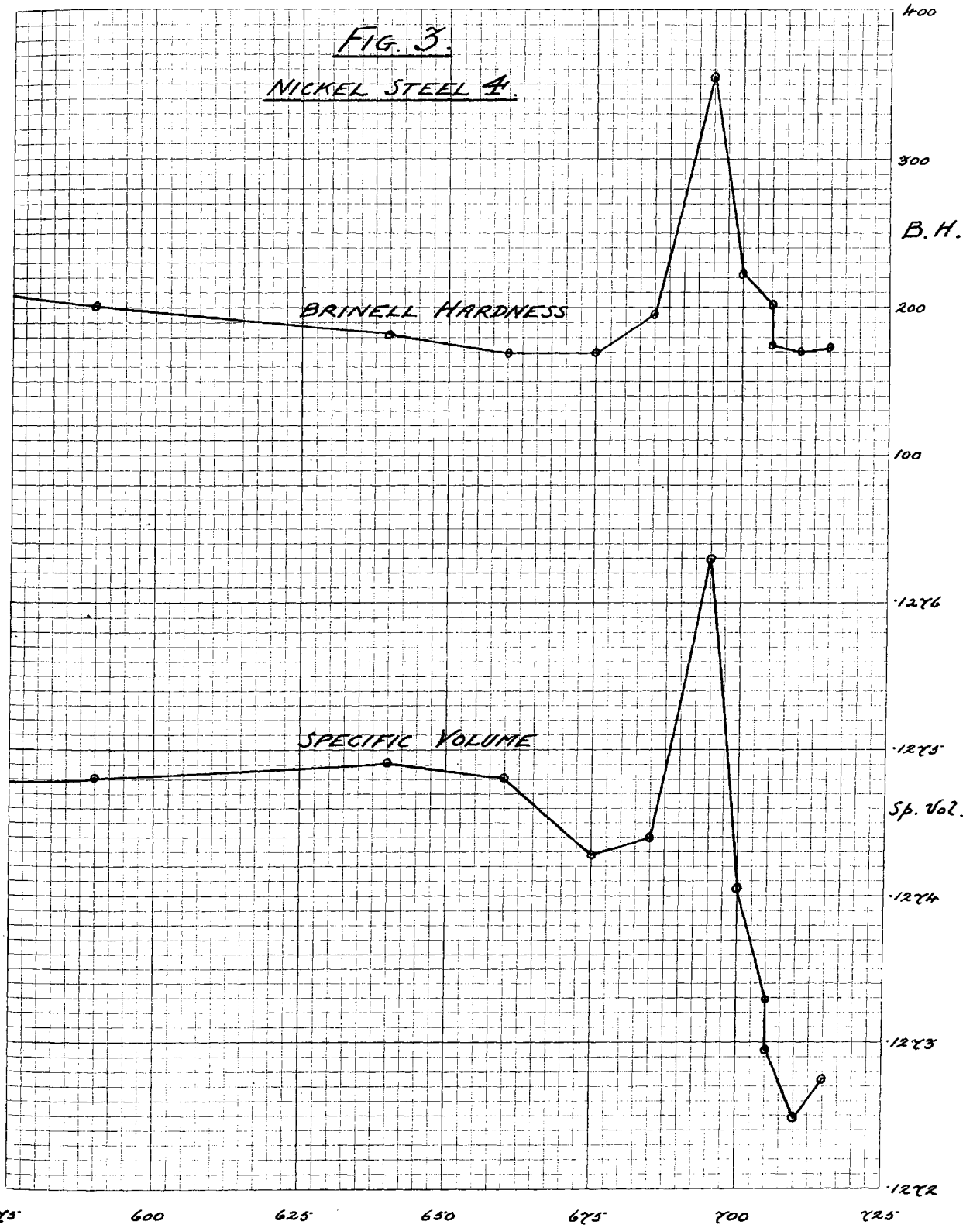


FIG. 3.
NICKEL STEEL 4.
BRINELL HARDNESS
SPECIFIC VOLUME
400
300
200
100
B. H.
·1276
·1275
·1274
·1273
·1272
Sp. Vol.
575
600
625
650
675
700
725
TEMPERATURE °C.

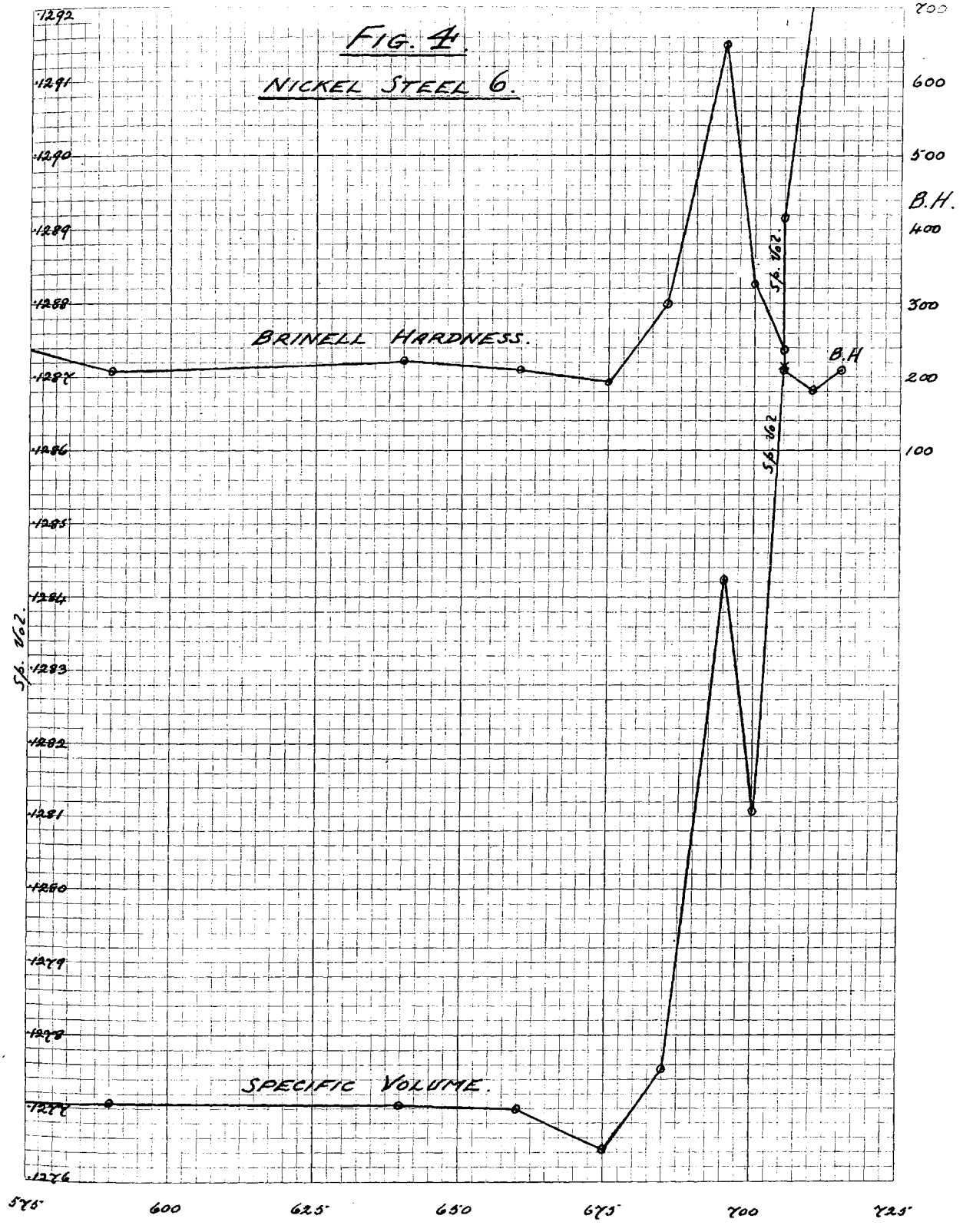


FIG. 4
NICKEL STEEL 6.
BRINELL HARDNESS.
SPECIFIC VOLUME.
SP. VOL.
1292
1291
1290
1289
1288
1287
1286
1285
1284
1283
1282
1281
1280
1279
1278
1277
1276
700
600
500
400
300
200
100
B.H.
SP. VOL.
B.H
575°
600
625°
650
675°
700
725°
TEMPERATURE °C.

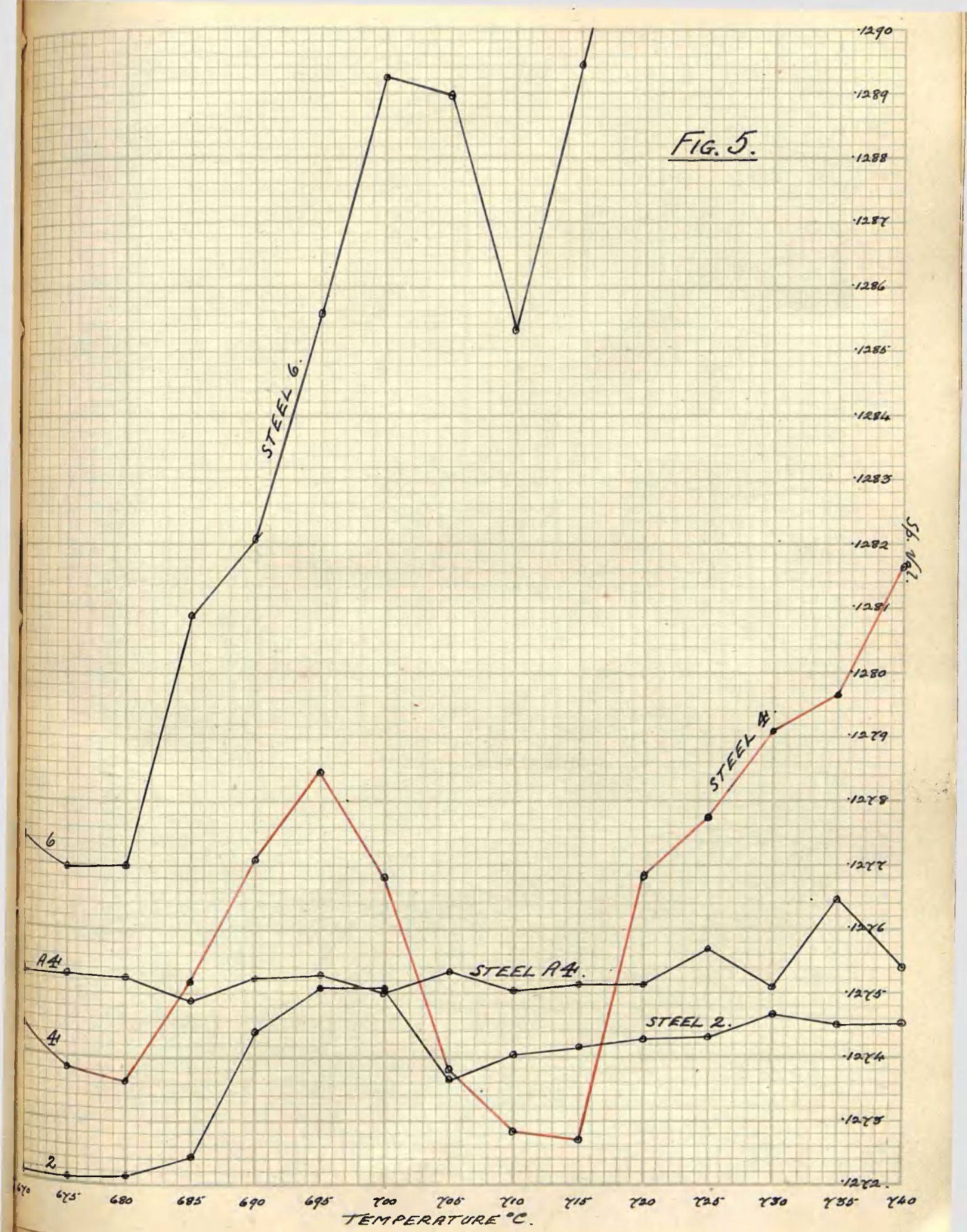


FIG. 5.
STEEL 6.
STEEL 4.
STEEL A4.
STEEL 2.
6
A4
4
2
Sp. Wt.
·1290
·1289
·1288
·1287
·1286
·1285
·1284
·1283
·1282
·1281
·1280
·1279
·1278
·1277
·1276
·1275
·1274
·1273
·1272
670
675
680
685
690
695
700
705
710
715
720
725
730
735
740
TEMPERATURE °C.

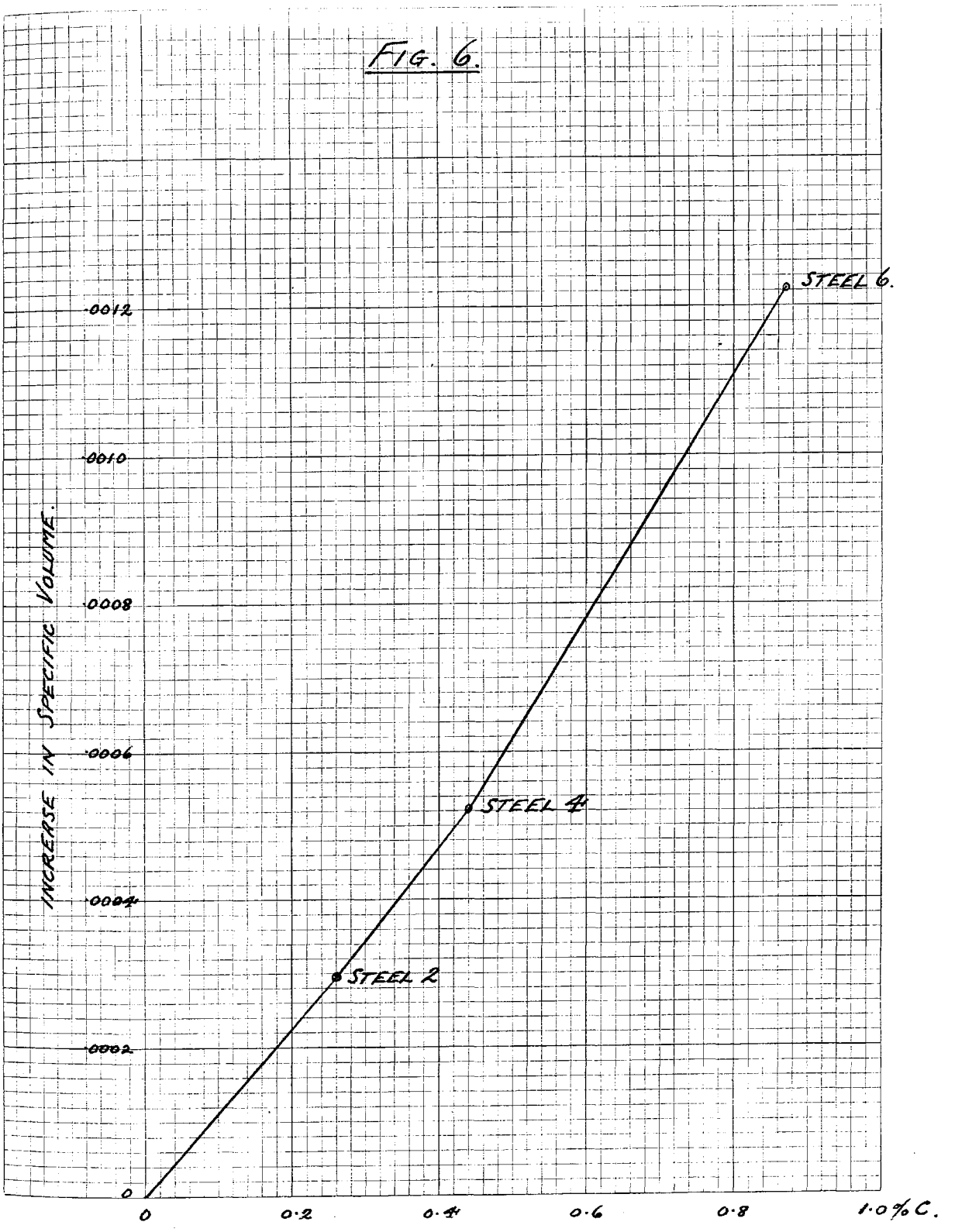


FIG. 6.
STEEL 6.
STEEL 4.
STEEL 2.
·0012
·0010
·0008
·0006
·0004
·0002
0
INCREASE IN SPECIFIC VOLUME.
0
0·2
0·4
0·6
0·8
1·0 % C.

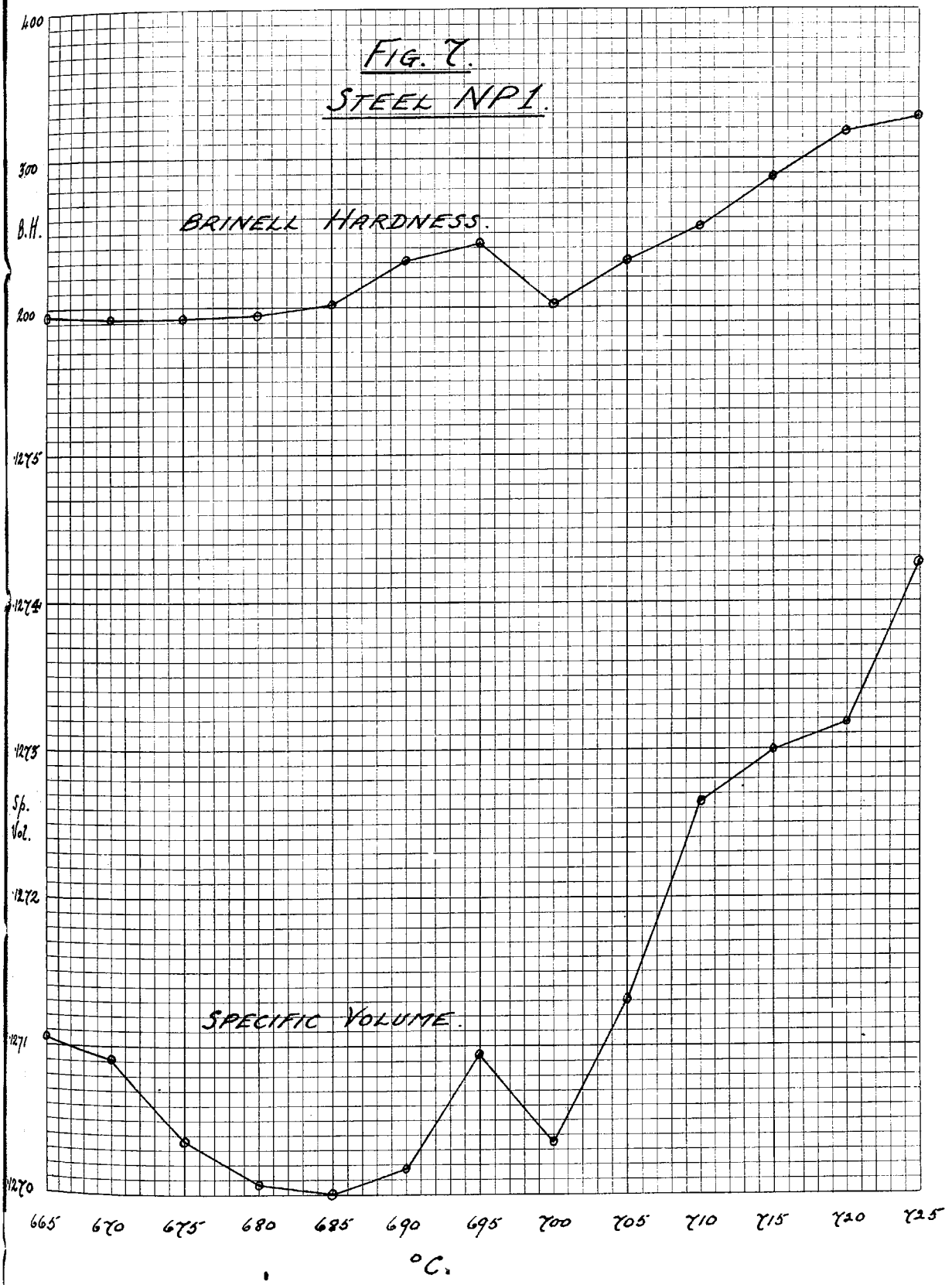


FIG. 7.
STEEL NP1.
BRINELL HARDNESS.
B.H.
400
300
200
SPECIFIC VOLUME.
Sp. Vol.
1275
1274
1273
1272
1271
1270
665
670
675
680
685
690
695
700
705
710
715
720
725
°C.

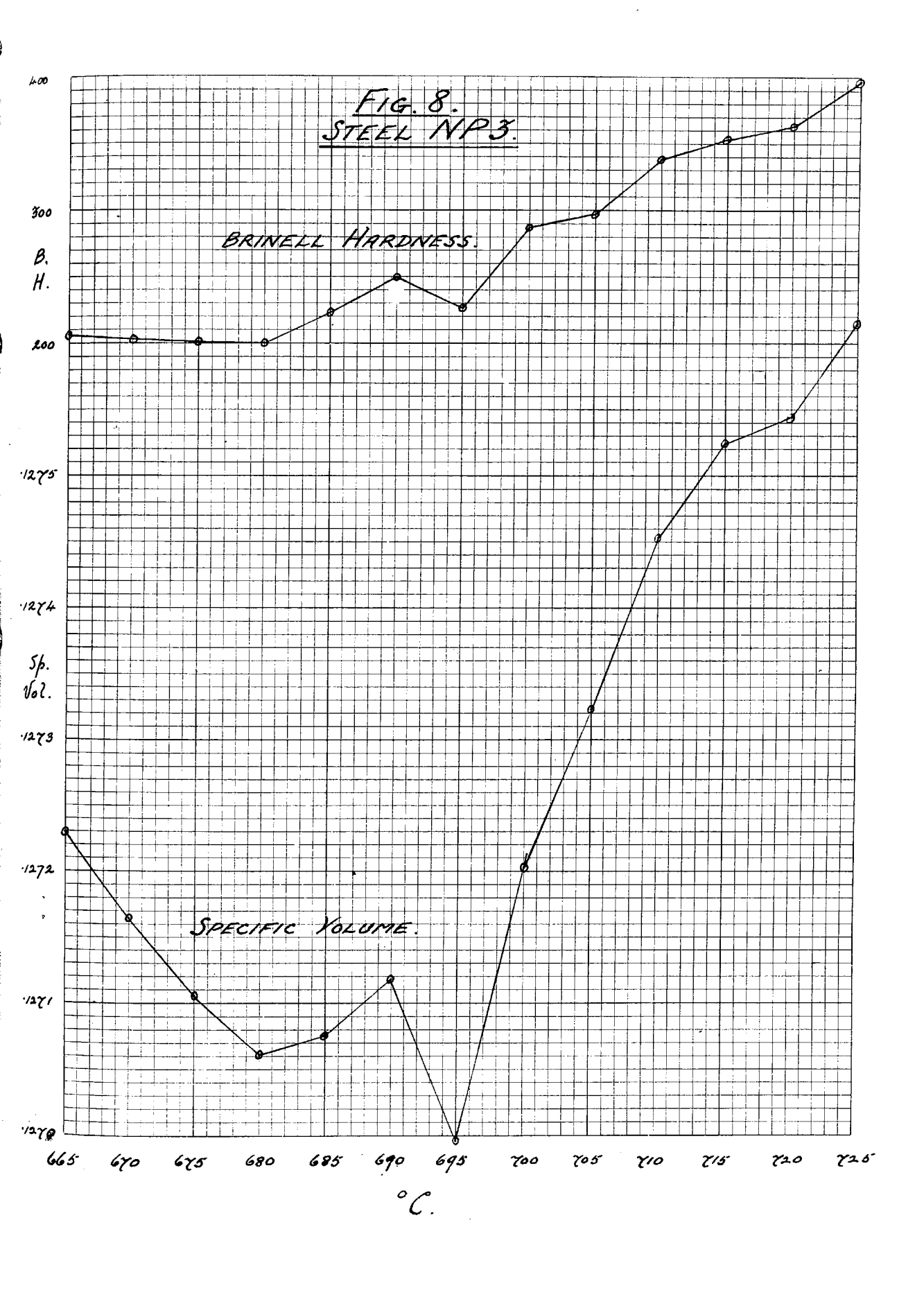


FIG. 8.
STEEL NP3.
BRINELL HARDNESS.
B.
H.
400
300
200
SPECIFIC VOLUME.
Sp.
Vol.
·1275
·1274
·1273
·1272
·1271
·1270
665
670
675
680
685
690
695
700
705
710
715
720
725
°C.

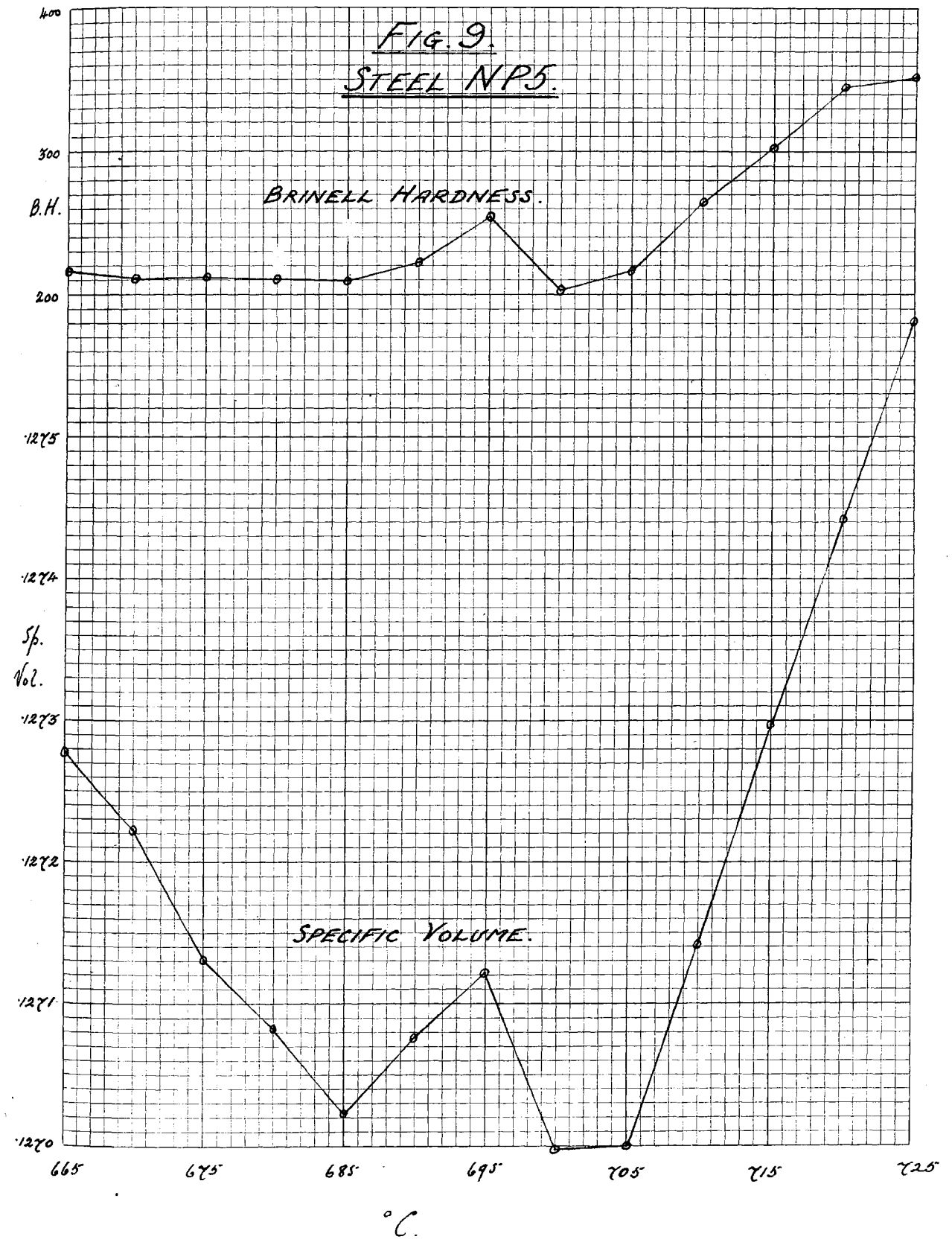


FIG. 9.
STEEL NP5.
BRINELL HARDNESS.
B.H.
400
300
200
SPECIFIC VOLUME.
Sp. Vol.
·1275
·1274
·1273
·1272
·1271
·1270
665
675
685
695
705
715
725
°C.

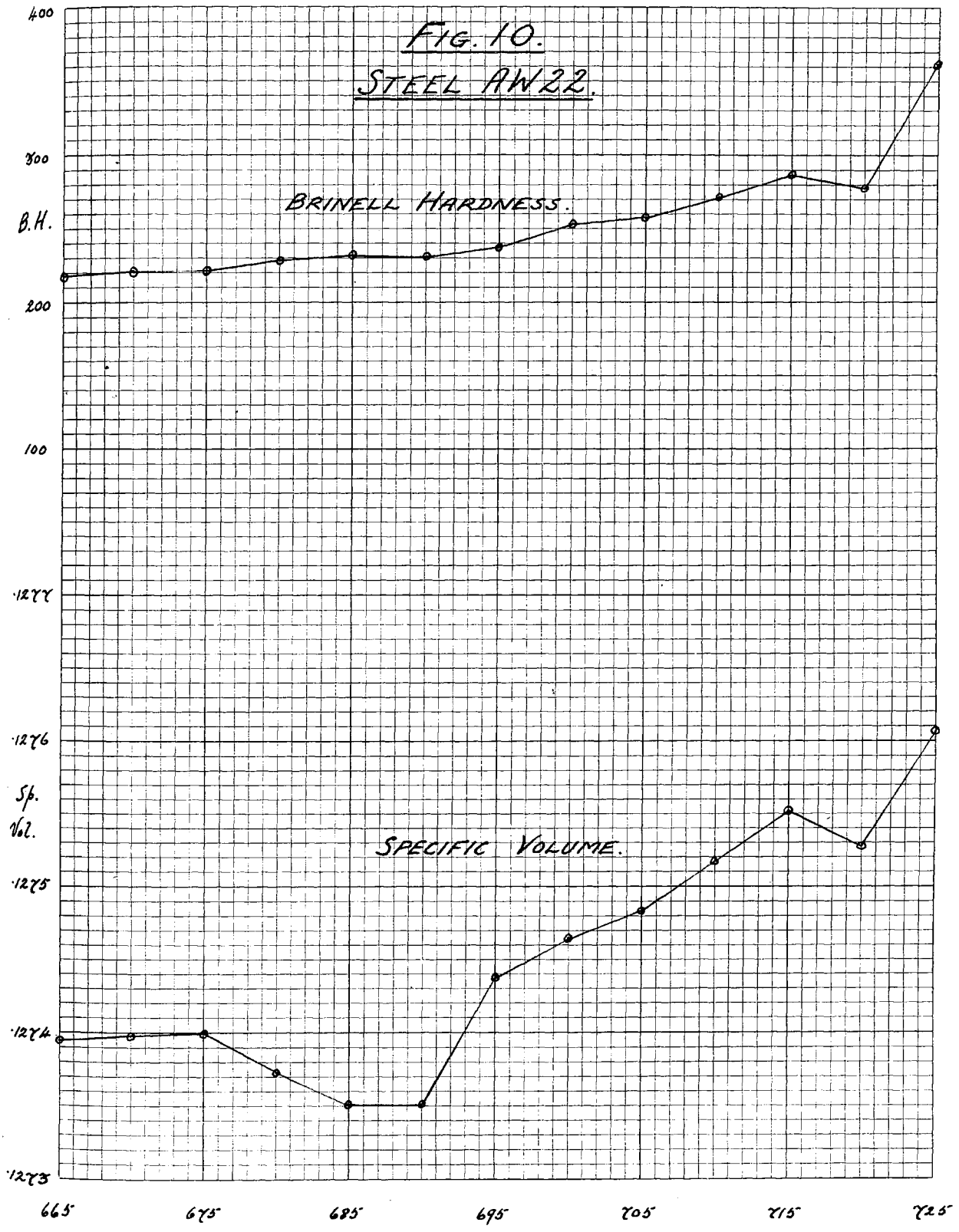


FIG. 10.
STEEL AW22.
400
300
B.H.
BRINELL HARDNESS.
200
100
·1277
·1276
Sp.
Vol.
SPECIFIC VOLUME.
·1275
·1274
·1273
665°
675°
685°
695°
705°
715°
725°
°C.

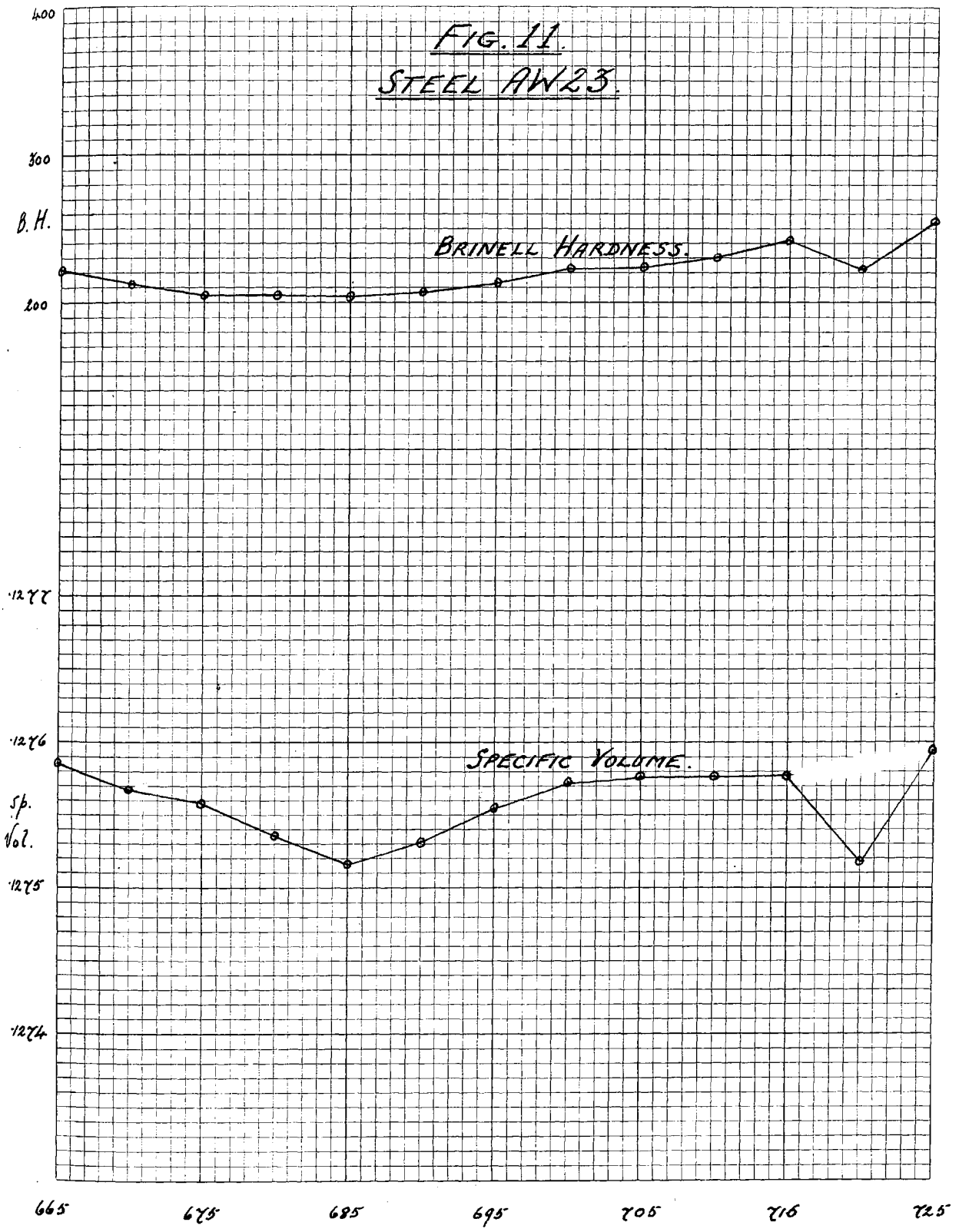
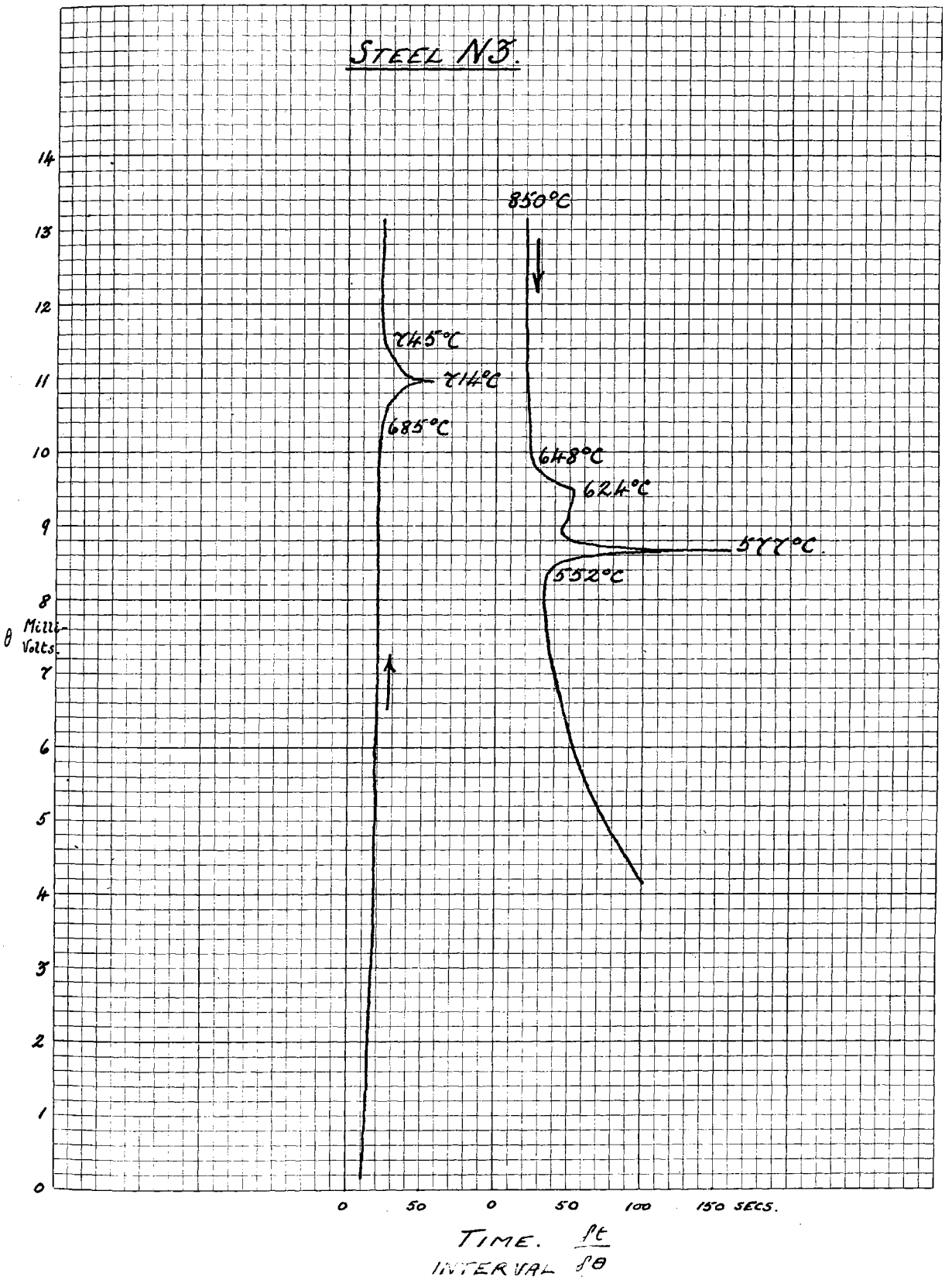


FIG. 11.
STEEL AW23.
400
300
B.H.
BRINELL HARDNESS.
200
·1277
·1276
SPECIFIC VOLUME.
Sp.
Vol.
·1275
·1274
665
675
685
695
705
716
725
°C.

FIG. 12.


STEEL N3.
850°C
745°C
714°C
685°C
648°C
624°C
577°C.
552°C
θ Milli-Volts
14
13
12
11
10
9
8
7
6
5
4
3
2
1
0
0
50
0
50
100
150 SECS.
TIME.
INTERVAL δt/δθ

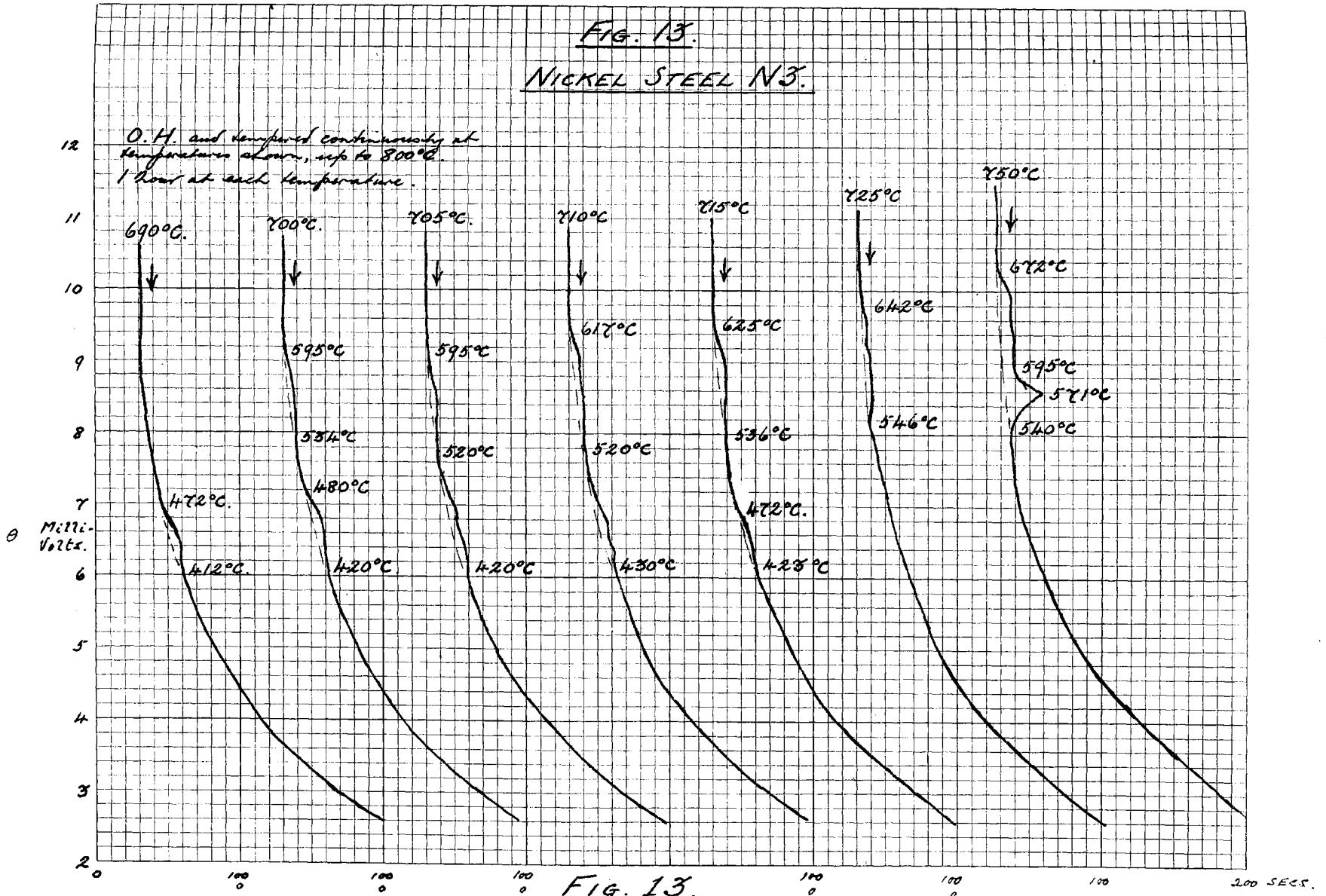


FIG. 13.
NICKEL STEEL N3.
O.H. and tempered continuously at temperatures shown, up to 800°C. 1 hour at each temperature.
690°C.
700°C.
705°C.
710°C
715°C
725°C
750°C
772°C
642°C
625°C
617°C
595°C
595°C
595°C
571°C
554°C
546°C
540°C
536°C
520°C
520°C
480°C
472°C.
472°C.
420°C
420°C
430°C
423°C
412°C.
θ Milli-Volts.
12
11
10
9
8
7
6
5
4
3
2
0
100
100
100
100
100
200 SECS.

FIG. 13.

dt/dθ.

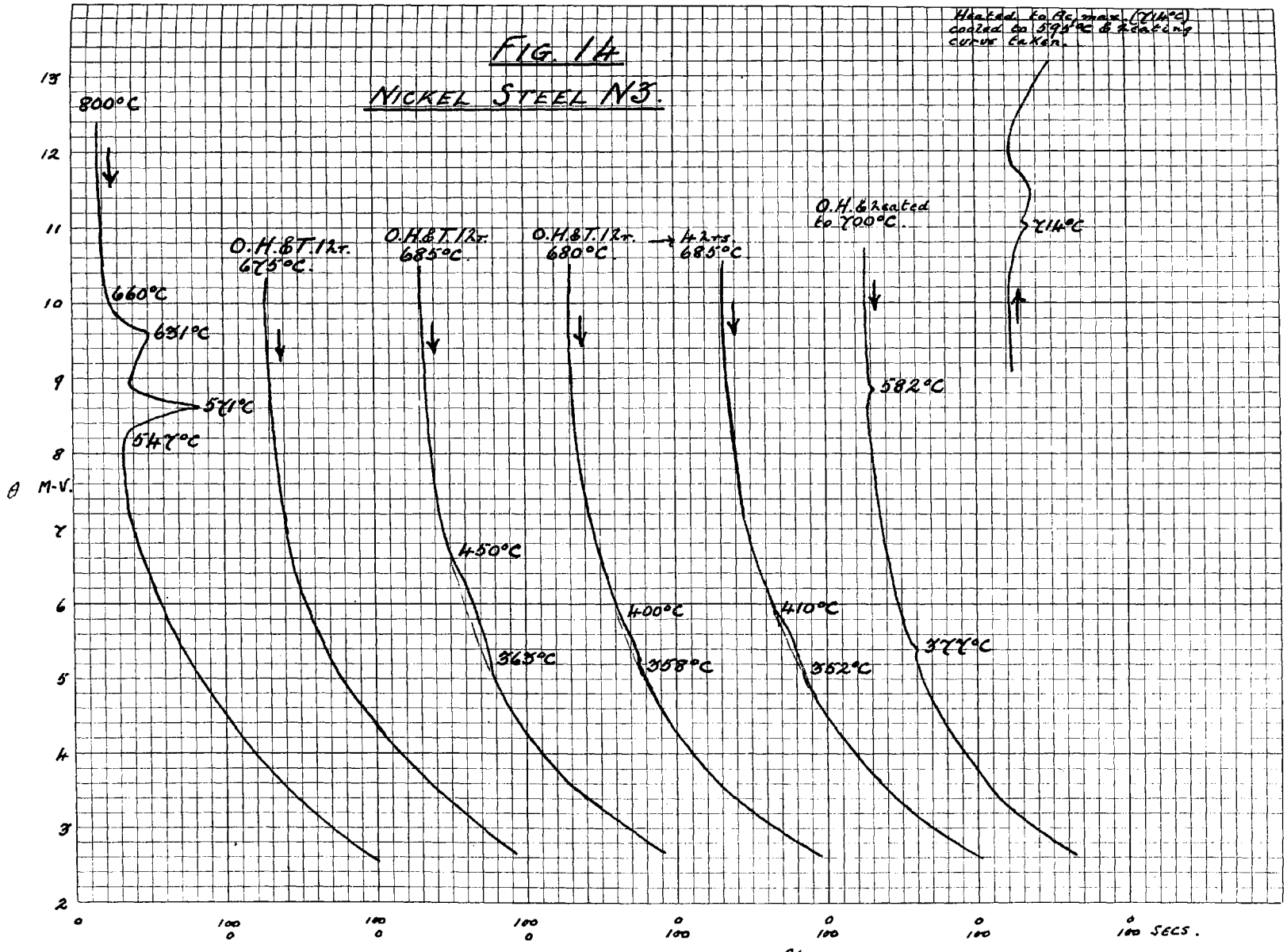
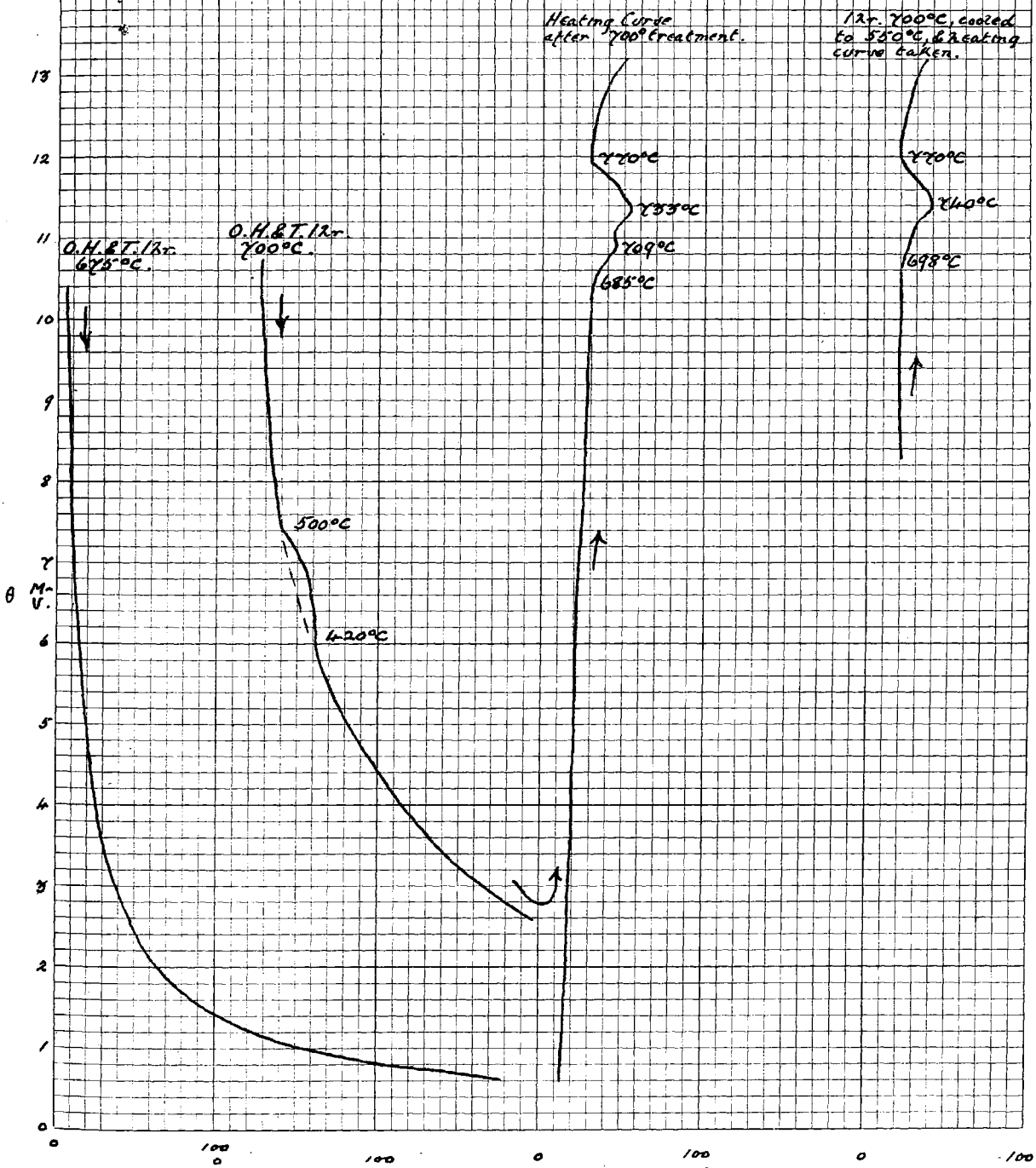


FIG. 14
NICKEL STEEL N3.
Heated to Ac max. (714°C) cooled to 594°C & heating curves taken.
800°C
660°C
631°C
571°C
547°C
O.H.&T. 12 hr. 675°C.
O.H.&T. 12 hr. 685°C.
O.H.&T. 12 hr. 680°C.
→ 42 hrs 685°C
O.H. & heated to 700°C.
714°C
582°C
450°C
400°C
410°C
377°C
365°C
358°C
352°C
θ M-V.
2
3
4
5
6
7
8
9
10
11
12
13
0
100
0
100
0
100
0
100
0
100
0
100 SECS.
TIME. INTERVAL, $\frac{\delta t}{\delta \theta}$.

FIG. 15.

NICKEL STEEL N3.



TIME. INTERVAL $\frac{\delta t}{\delta \theta}$.

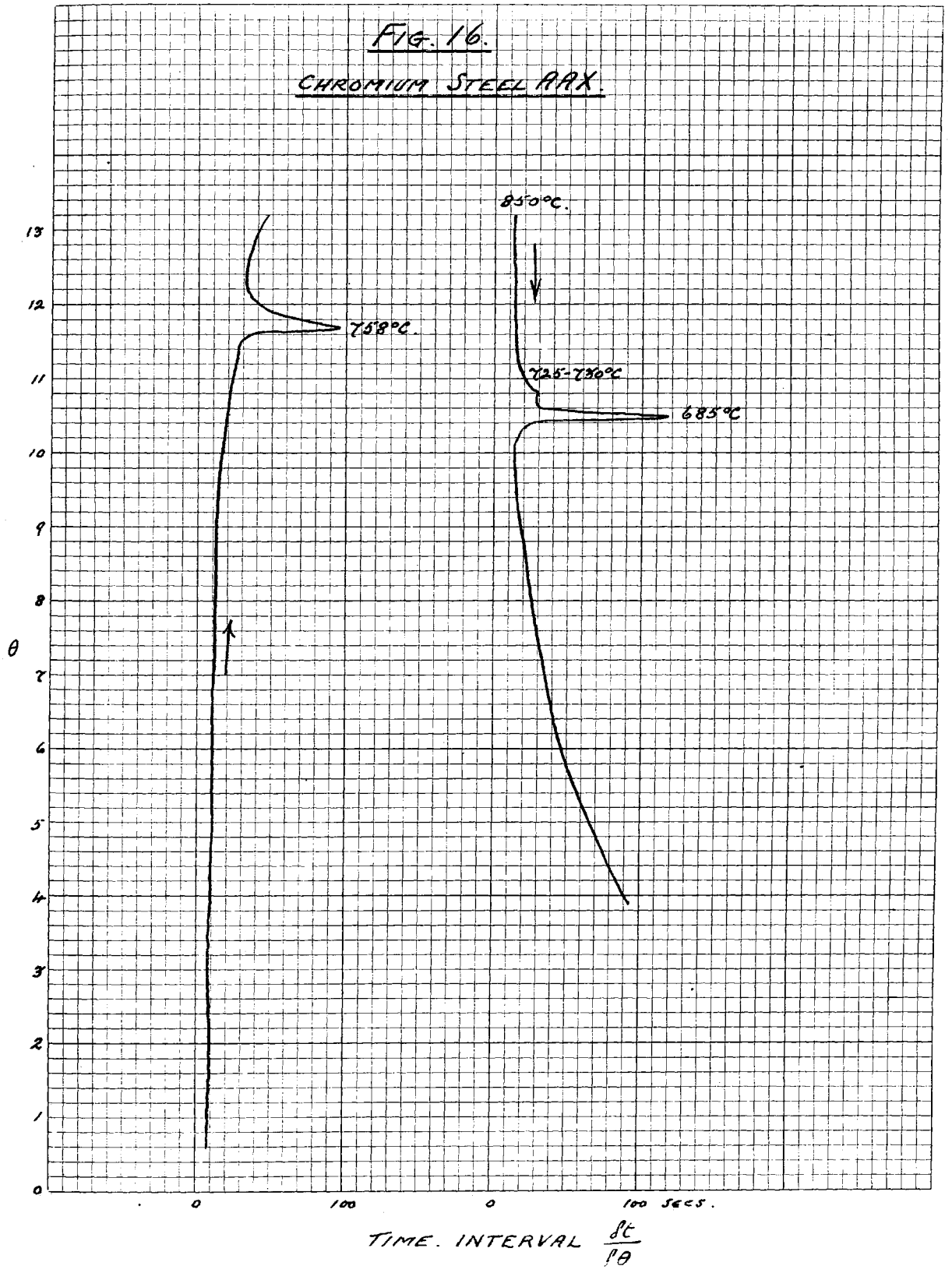


FIG. 16.
CHROMIUM STEEL AAX.
850°C.
758°C.
725-730°C
685°C
θ
13
12
11
10
9
8
7
6
5
4
3
2
1
0
0
100
0
100 SECS.
TIME. INTERVAL δt/δθ

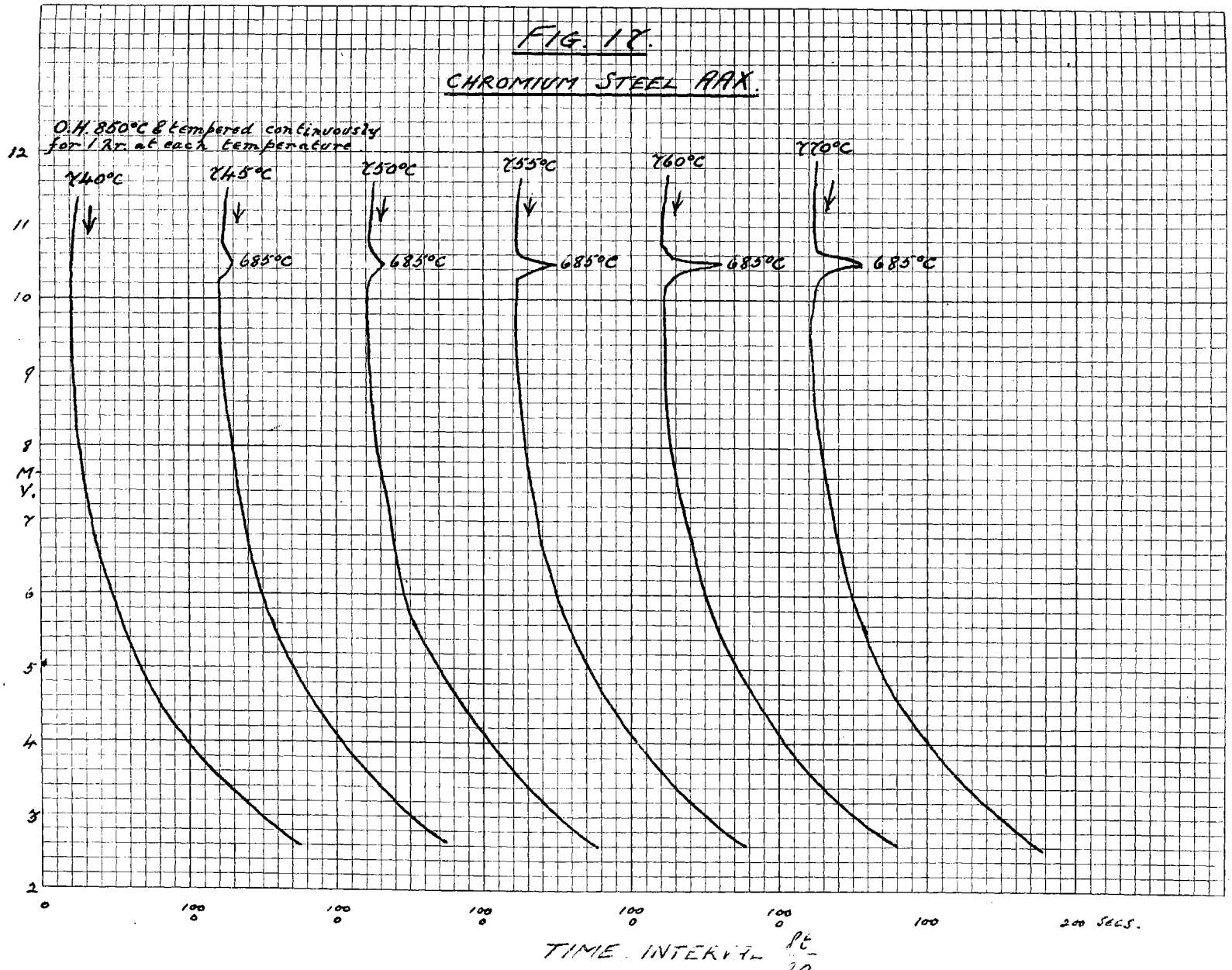
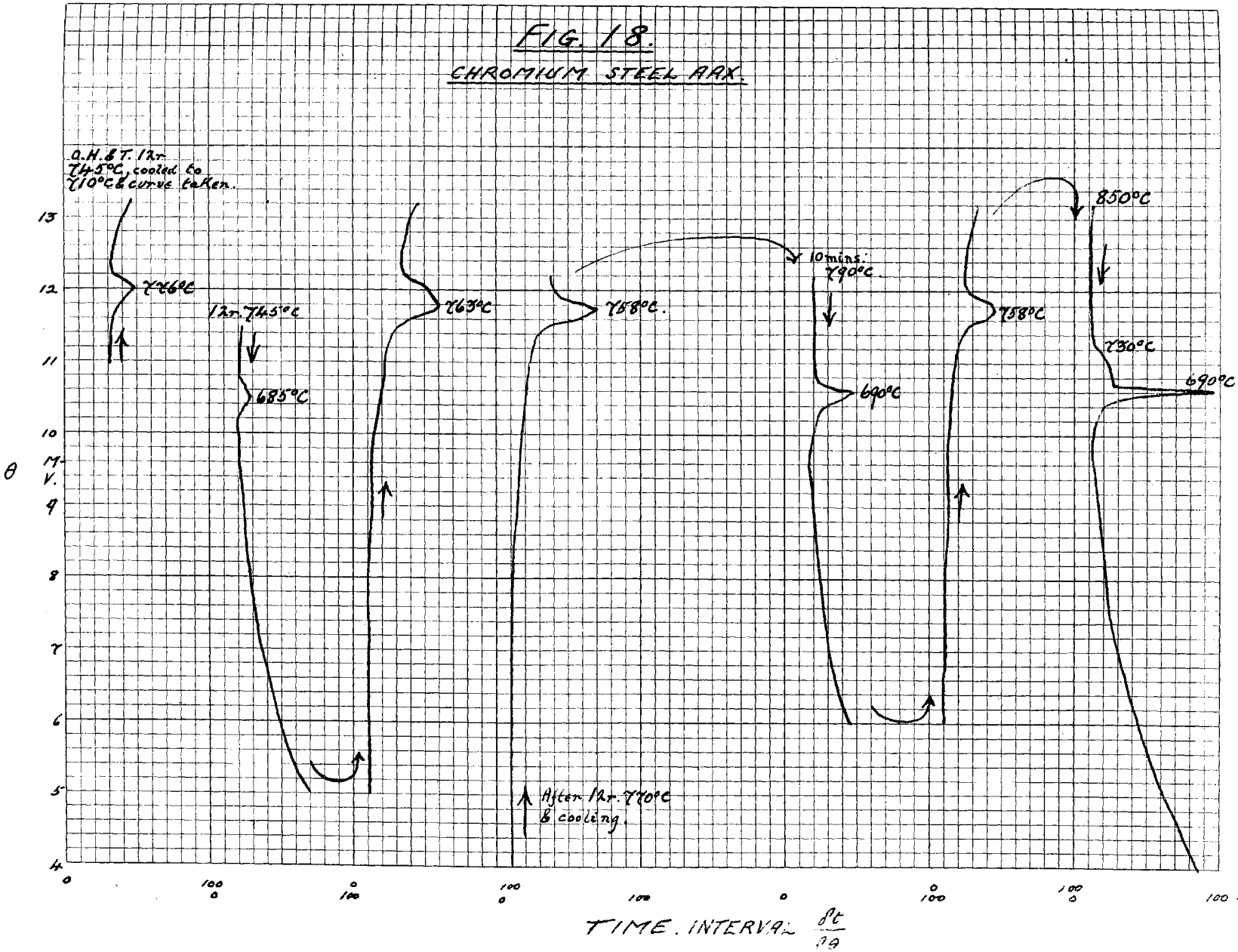


FIG. 12.
CHROMIUM STEEL AAX.
O.H. 850°C & tempered continuously for 1 hr. at each temperature.
740°C
745°C
750°C
755°C
760°C
770°C
685°C
685°C
685°C
685°C
685°C
12
11
10
9
8
M.V.
7
6
5
4
3
2
0
100
100
100
100
100
100
200 SECS.
TIME INTERVAL

FIG. 18.

CHROMIUM STEEL AAX.



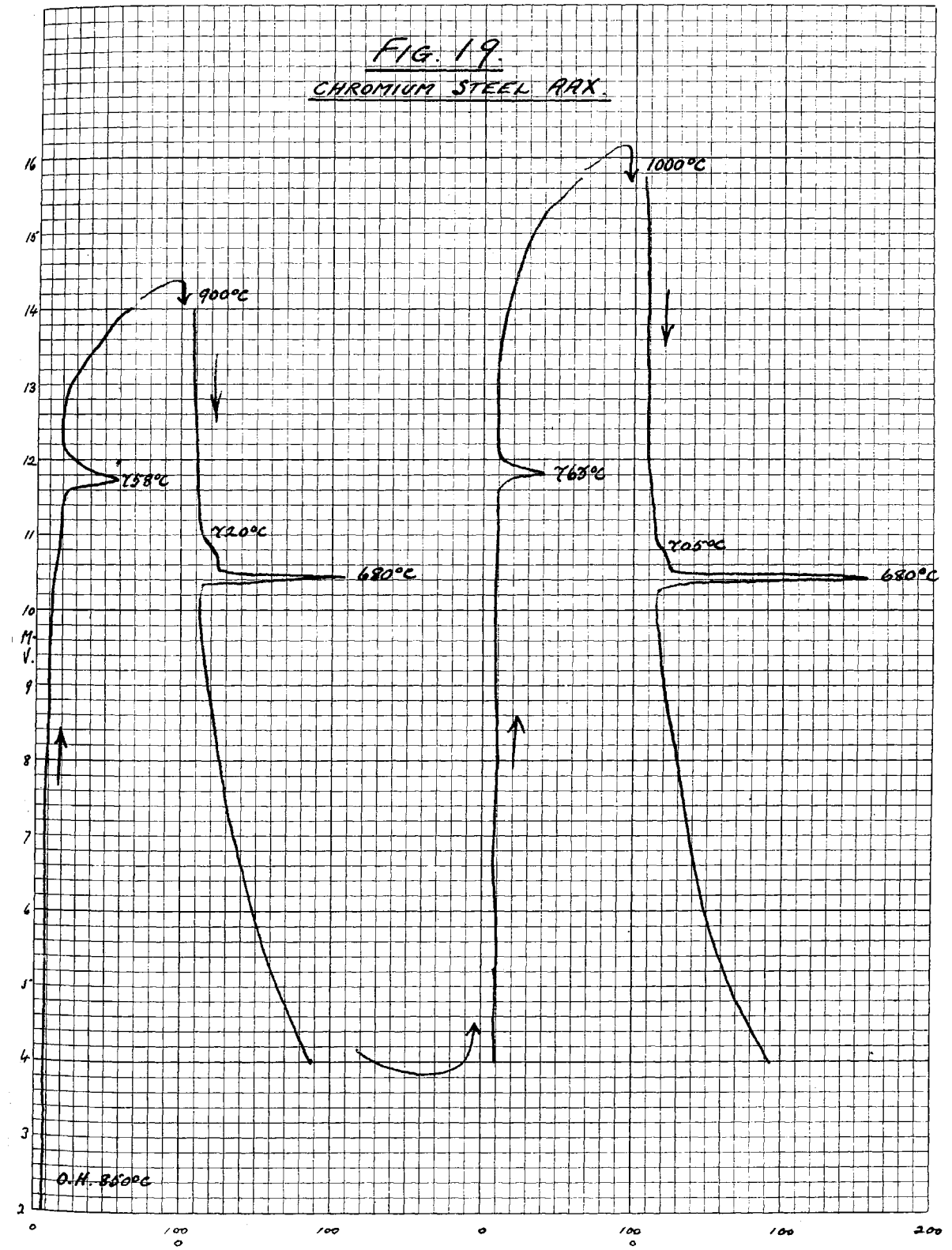


FIG. 19.
CHROMIUM STEEL RRX.
1000°C
900°C
763°C
758°C
720°C
705°C
680°C
680°C
M.V.
O.H. 850°C
16
15
14
13
12
11
10
9
8
7
6
5
4
3
2
0
100
100
0
100
100
200 SECS.
TIME.
INTERVAL.

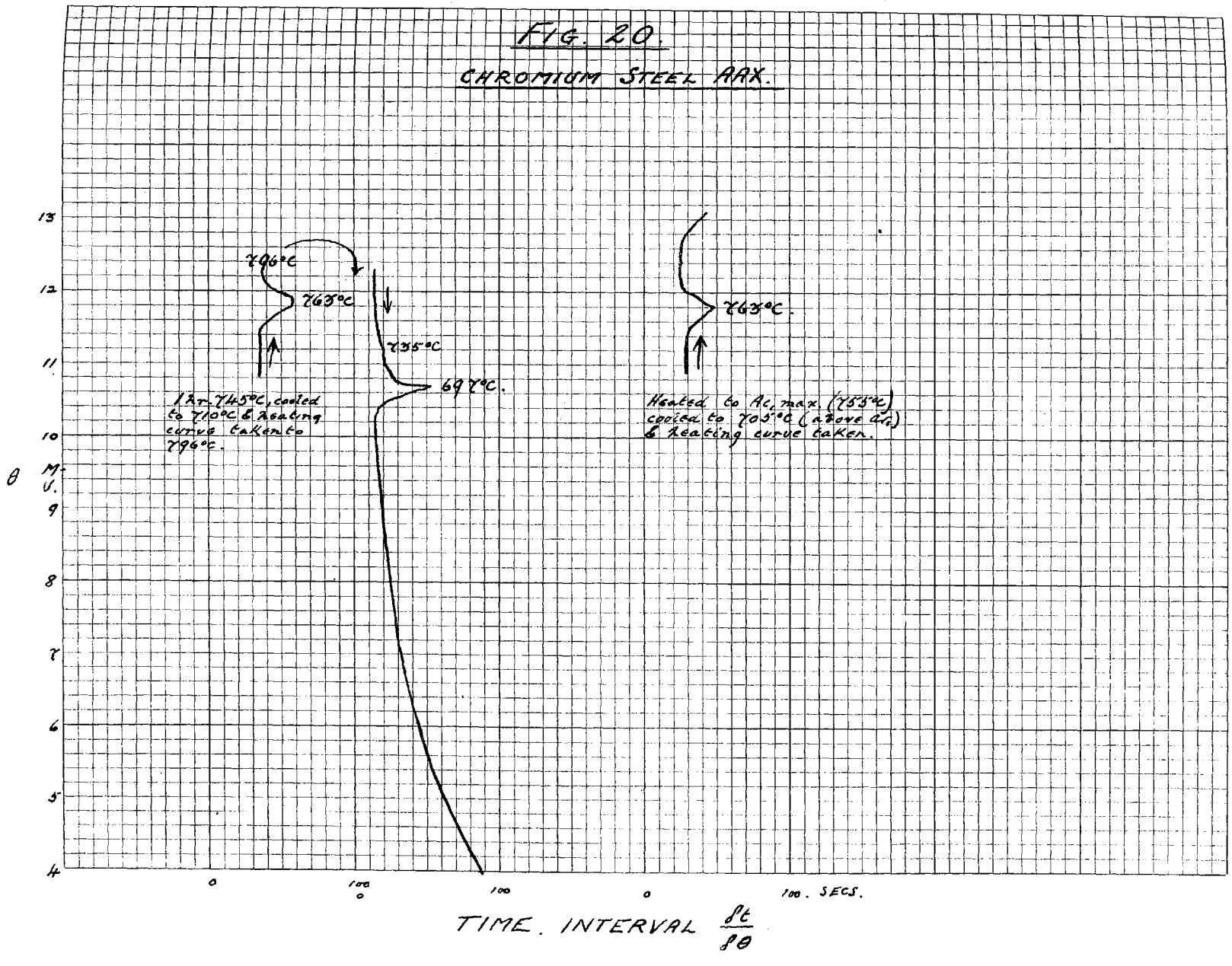
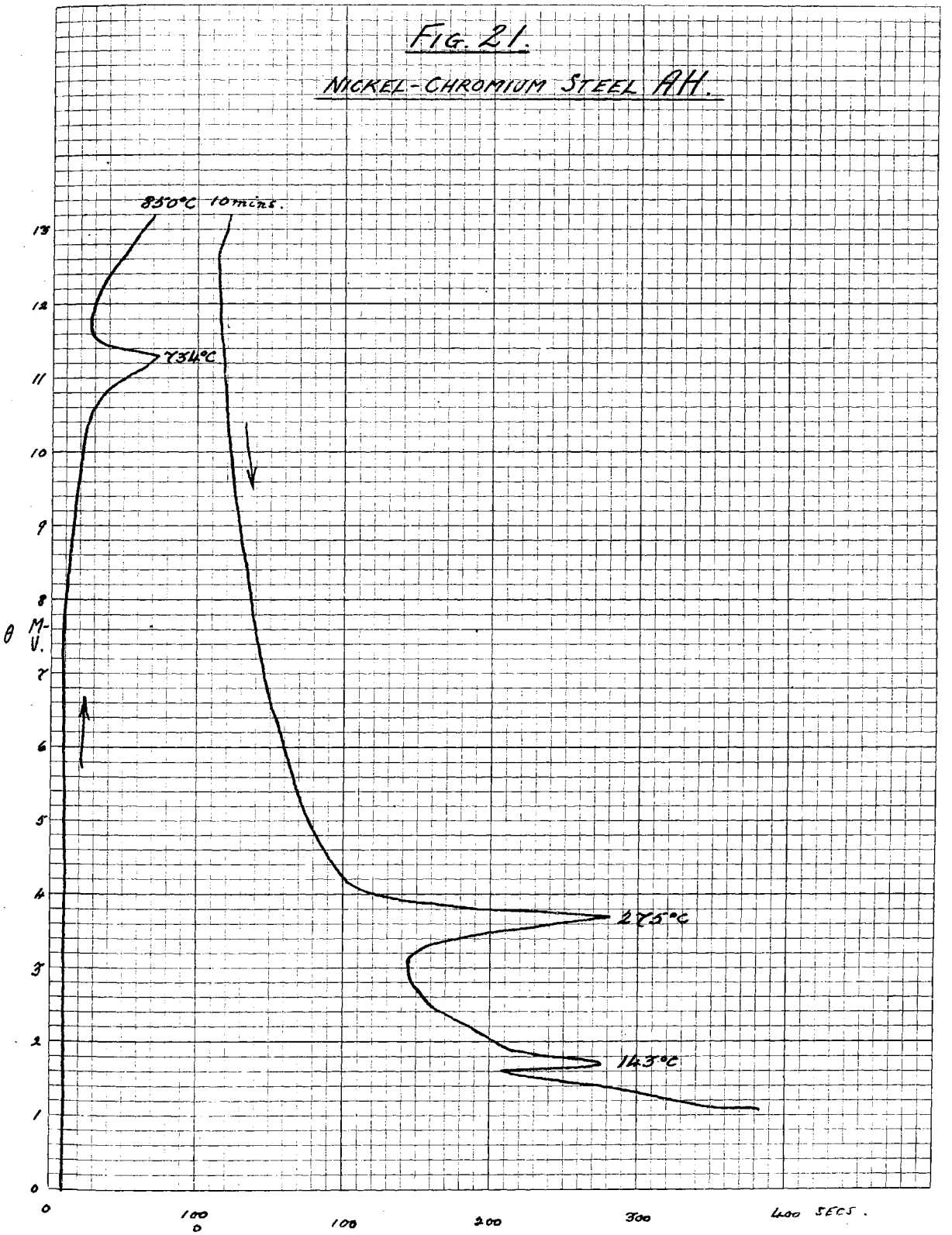


FIG. 20.
CHROMIUM STEEL AAX.
796°C
763°C
735°C
697°C.
Ht. 745°C, cooled to 710°C & heating curve taken to 796°C.
763°C.
Heated to Ac1 max. (755°C) cooled to 705°C (above Ar1) & heating curve taken.
θ
M.V.
13
12
11
10
9
8
7
6
5
4
0
100
0
100
0
100. SECS.
TIME. INTERVAL δt/δθ

FIG. 21.

NICKEL-CHROMIUM STEEL AH.



TIME. 2½
INTERVAL 2½

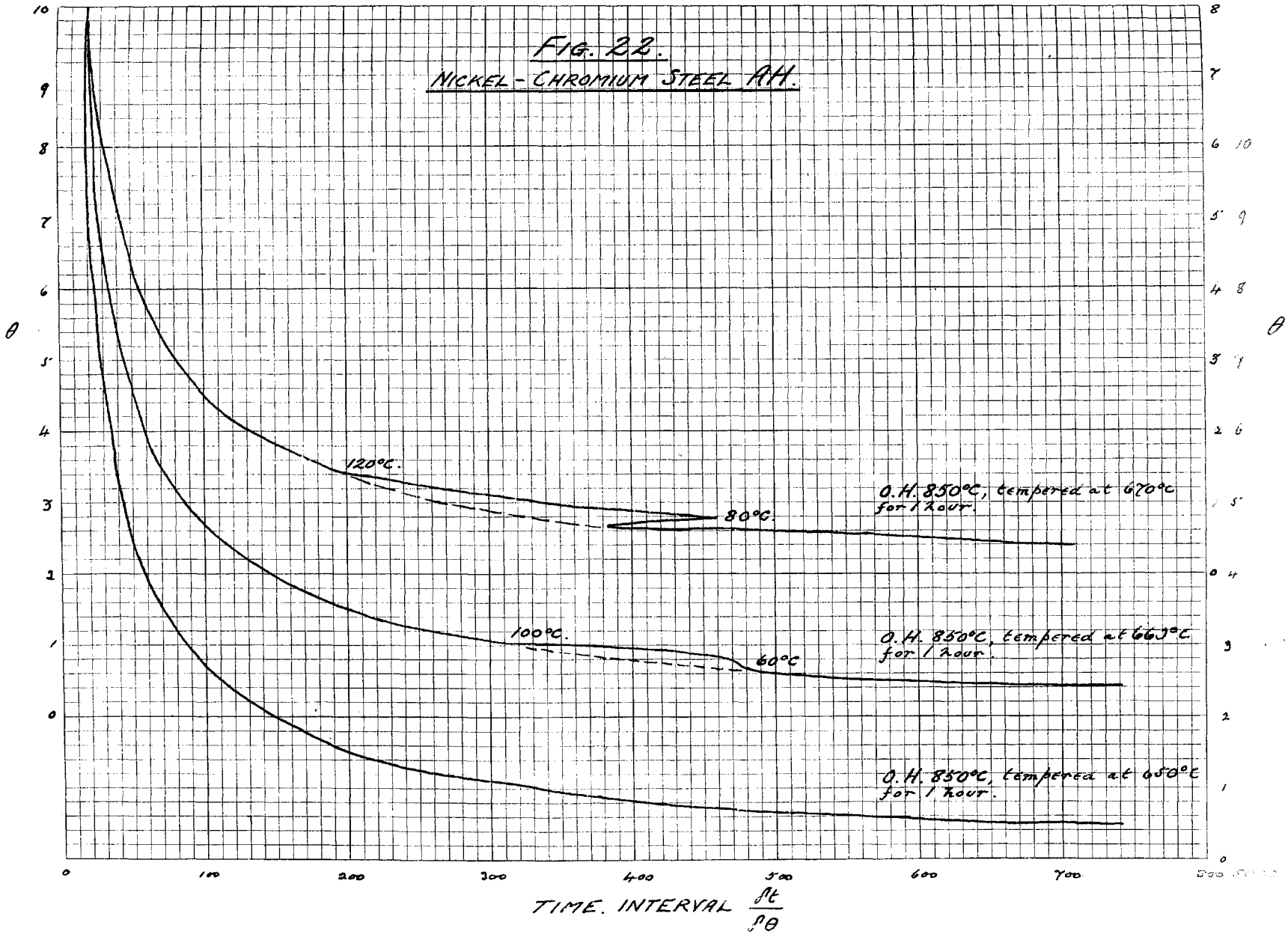


FIG. 22.
NICKEL-CHROMIUM STEEL AH.
θ
120°C.
80°C.
O.H. 850°C, tempered at 670°C for 1 hour.
100°C.
60°C
O.H. 850°C, tempered at 660°C for 1 hour.
O.H. 850°C, tempered at 650°C for 1 hour.
0
100
200
300
400
500
600
700
TIME. INTERVAL $\frac{\delta t}{\delta \theta}$

FIG. 23.

NICKEL-CHROMIUM STEEL NC.

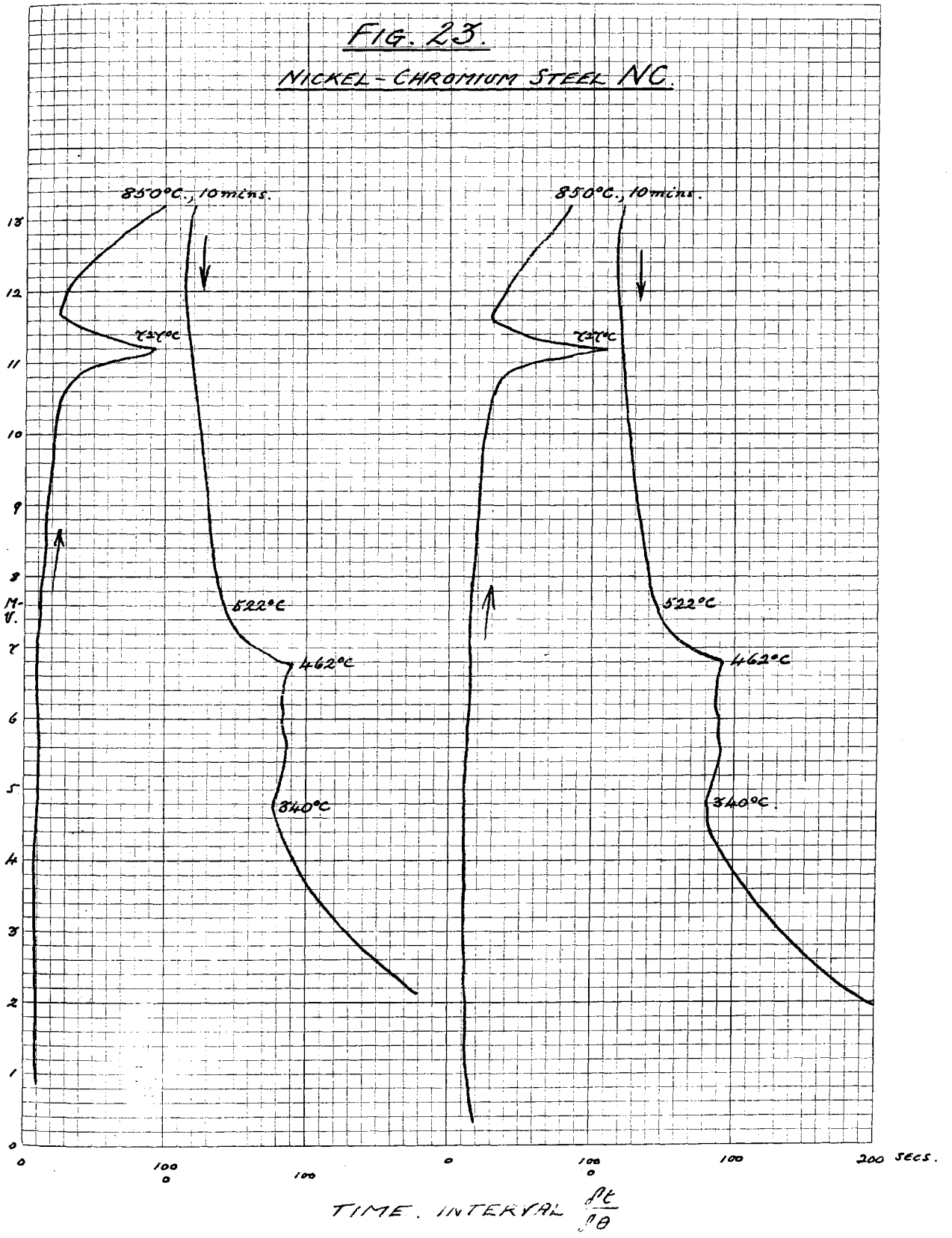
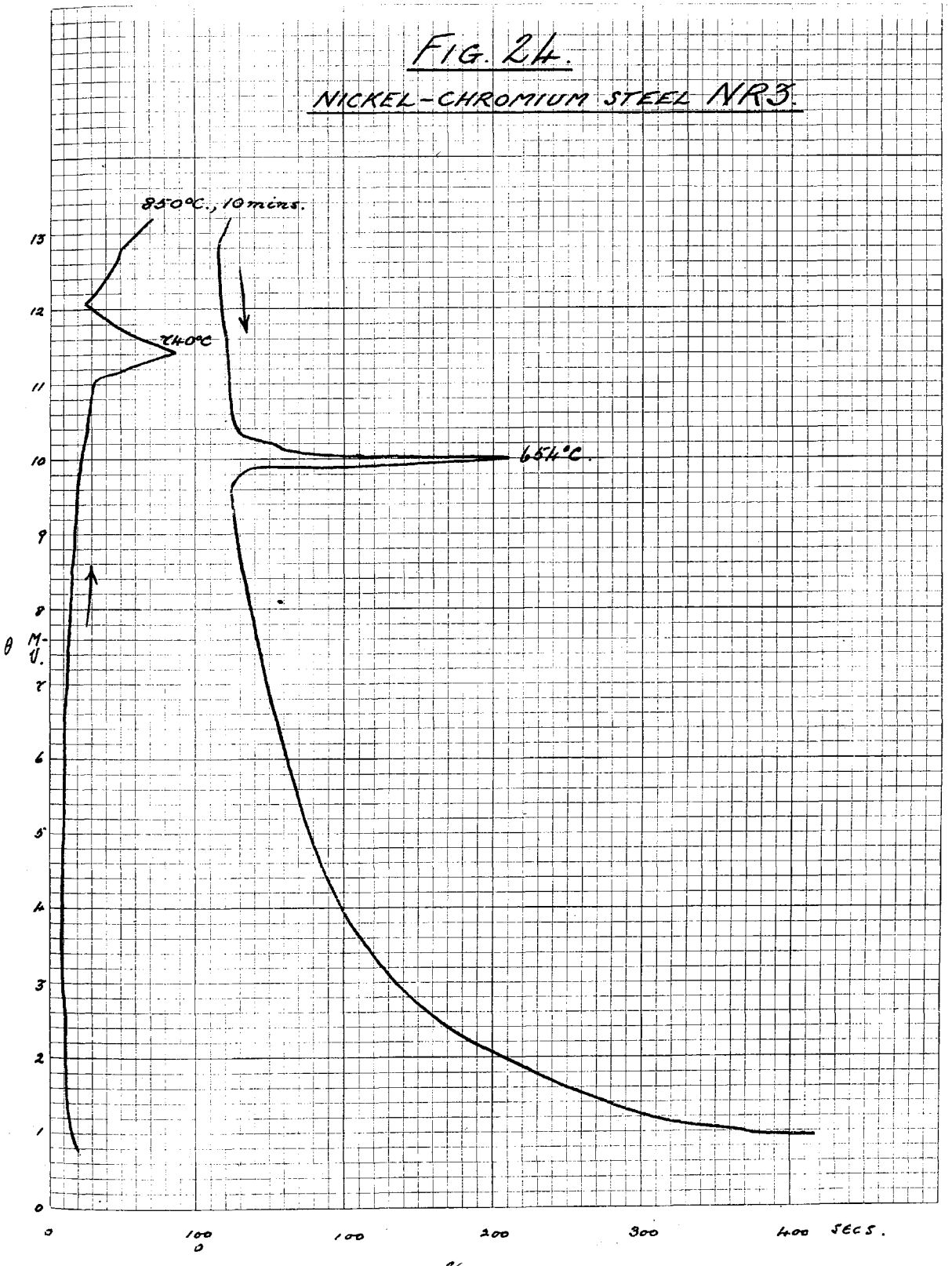


FIG. 24.

NICKEL-CHROMIUM STEEL NR3.



$\frac{dt}{d\theta}$

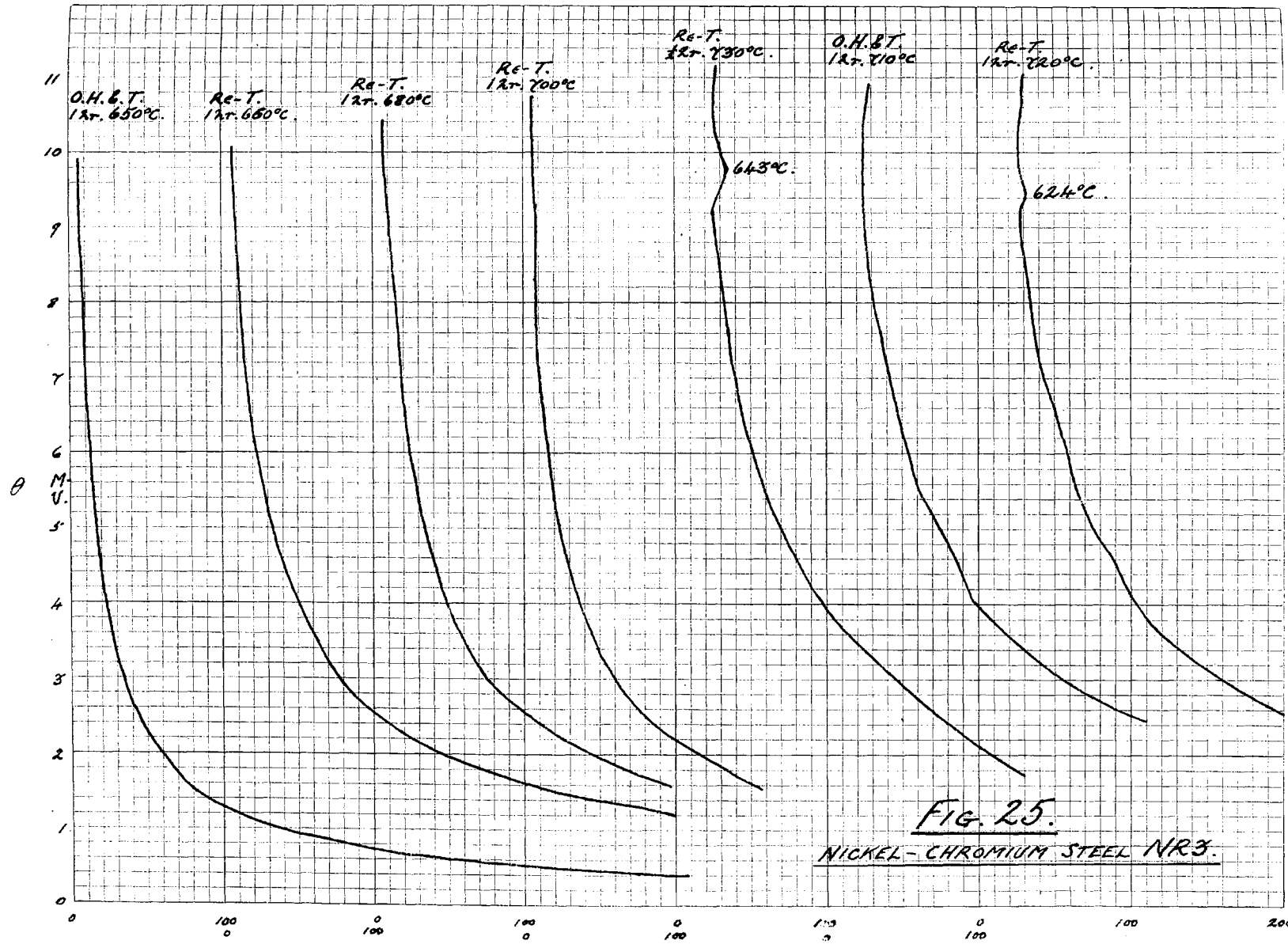
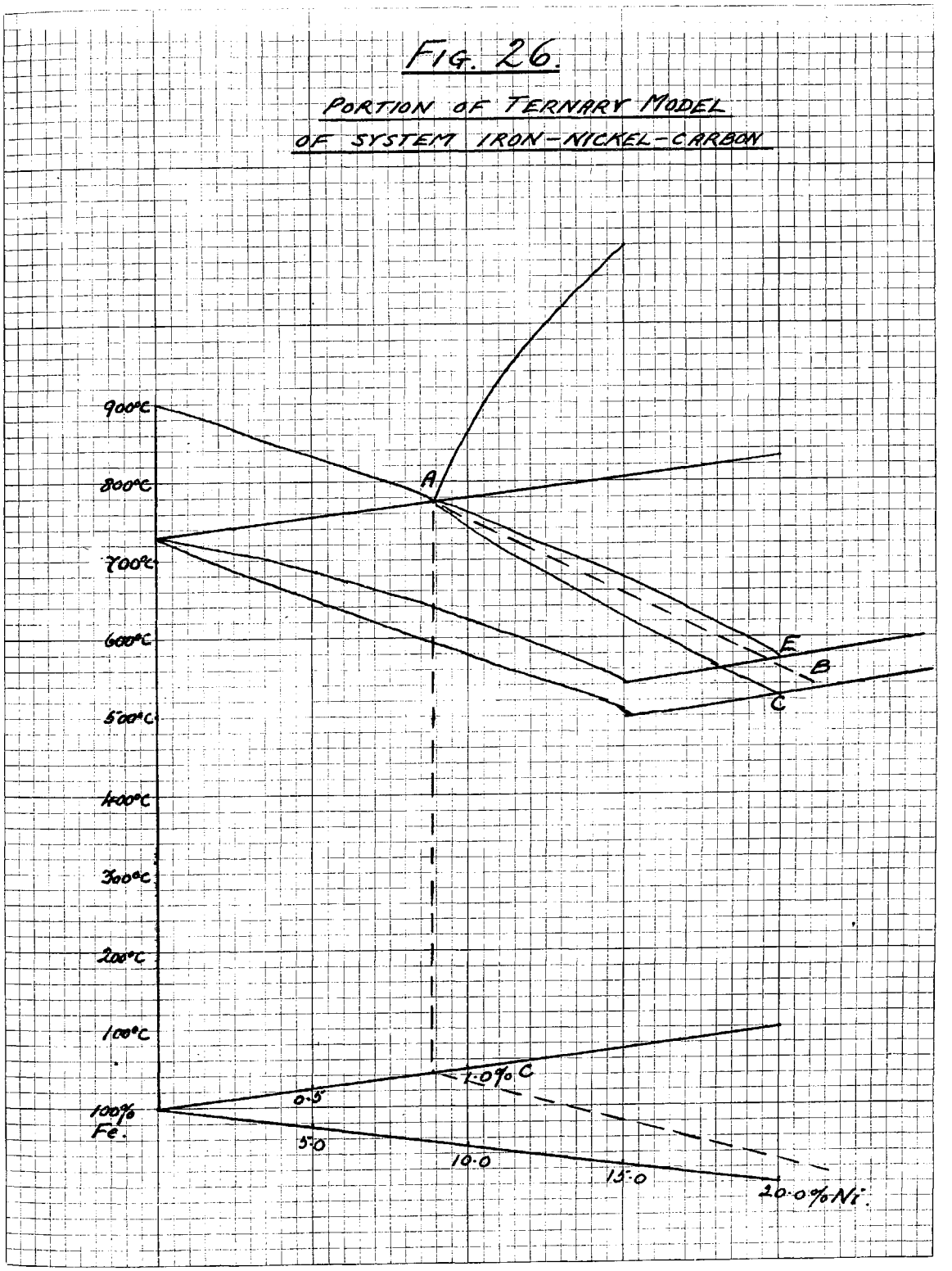


FIG. 25.

NICKEL-CHROMIUM STEEL NR3.

FIG. 26.

PORTION OF TERNARY MODEL OF SYSTEM IRON-NICKEL-CARBON



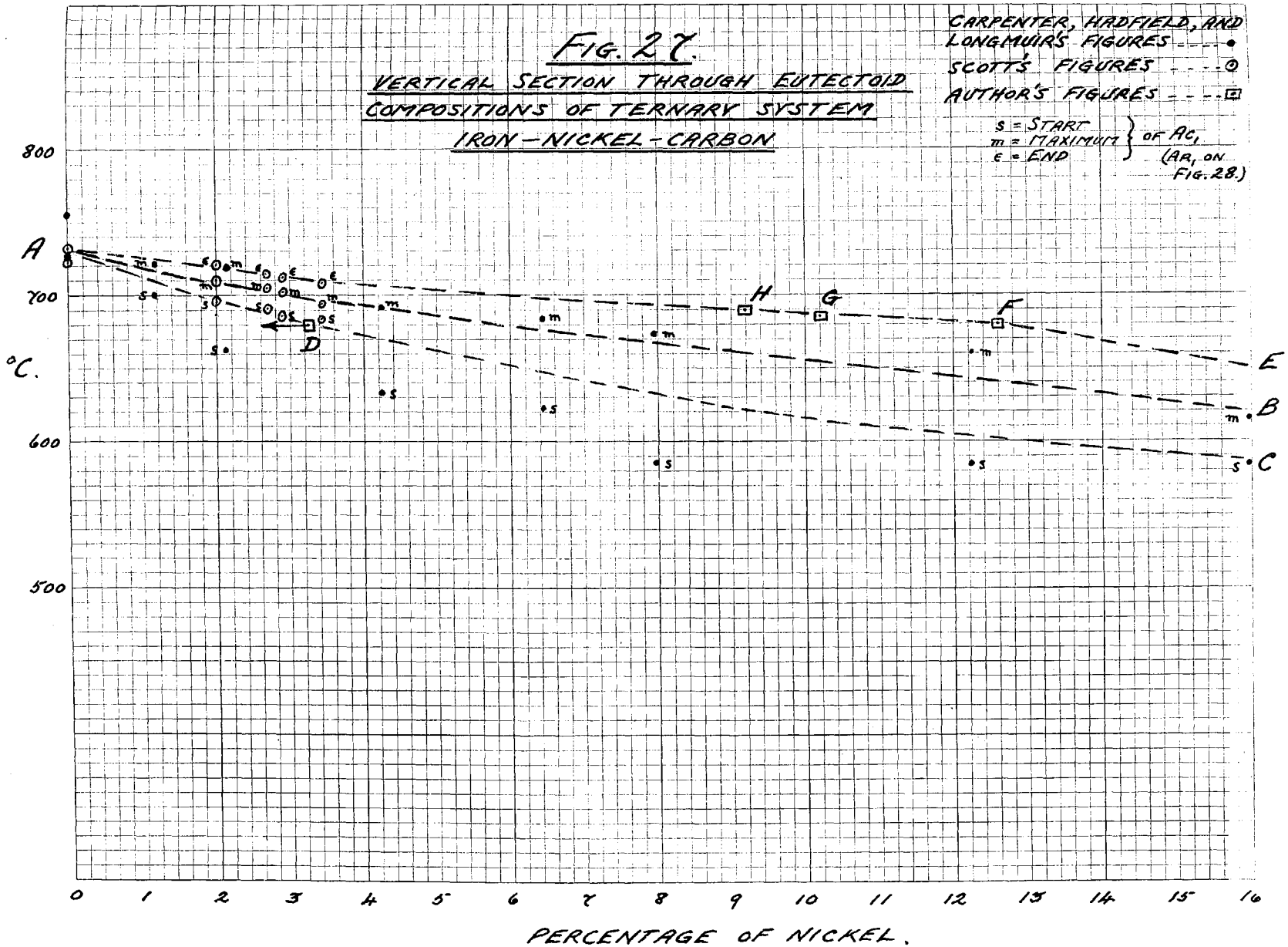


FIG. 27.
VERTICAL SECTION THROUGH EUTECTOID COMPOSITIONS OF TERNARY SYSTEM IRON–NICKEL–CARBON
CARPENTER, HADFIELD, AND LONGMUIR'S FIGURES
SCOTT'S FIGURES
AUTHOR'S FIGURES
s = START
m = MAXIMUM
ε = END
} OF Ac1
(Ar1 ON FIG. 28.)
800
700
600
500
°C.
A
H
G
F
D
E
B
C
0 1 2 3 4 5 6 7 8 9 10 11 12 13 14 15 16
PERCENTAGE OF NICKEL.

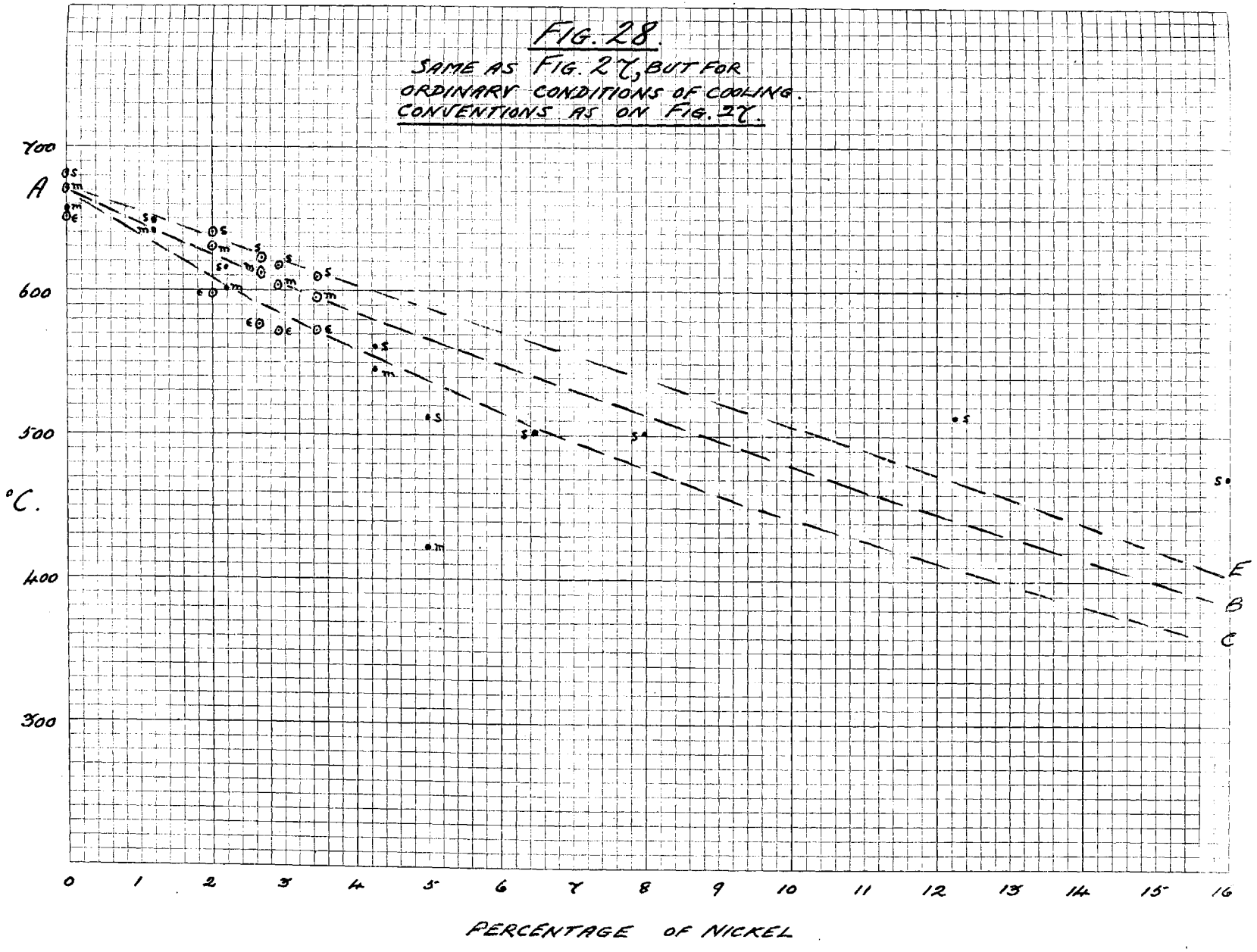


FIG. 28.
SAME AS FIG. 27, BUT FOR ORDINARY CONDITIONS OF COOLING.
CONVENTIONS AS ON FIG. 27.
700
A
600
500
°C.
400
300
0 1 2 3 4 5 6 7 8 9 10 11 12 13 14 15 16
PERCENTAGE OF NICKEL
E
B
C

Micro. 1.

Steel No. 6. x 600. 12 Hours at 685°C.
Boiled in Sodium Picrate Soln., then etched 1% HNO_3 in Alc.

Micro. 2.

Steel No. 4. x 600. 12 Hours at 695°C.
Etched with 1% HNO_3 in Alcohol.

Micro. 3.

Steel No.2. x 600.
12 Hours at 700°C.

Micro. 4.

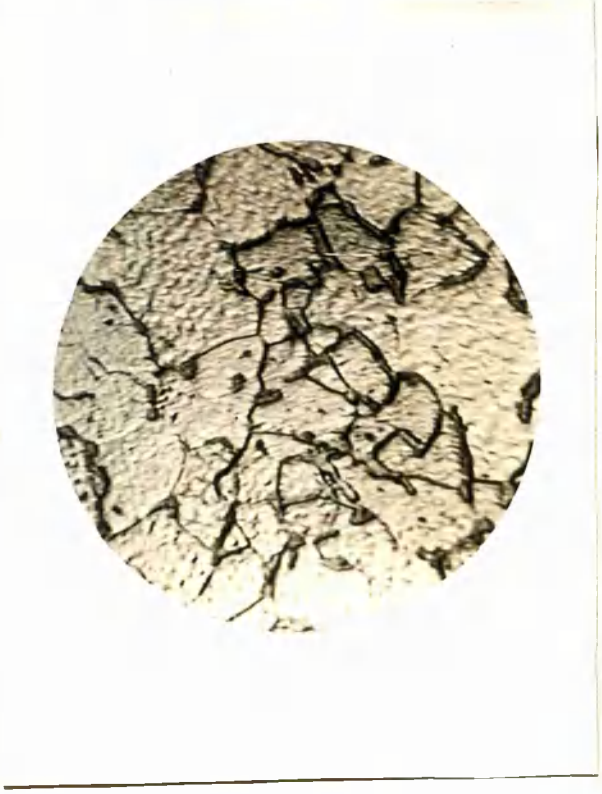
Steel No.2. x 600.
12 Hours at 705°C.

ALL ETCHED IN 1% HNO_3 IN ALCOHOL.

Micro. 5.

Steel No. 2. x 600.
24 Hours at 705°C.

Micro. 6.



Steel No. 2. x 600.
12 Hours at 710°C.

Micro. 7.

Steel No.2. x 600.
12 Hours at715°C.

Micro. 8.

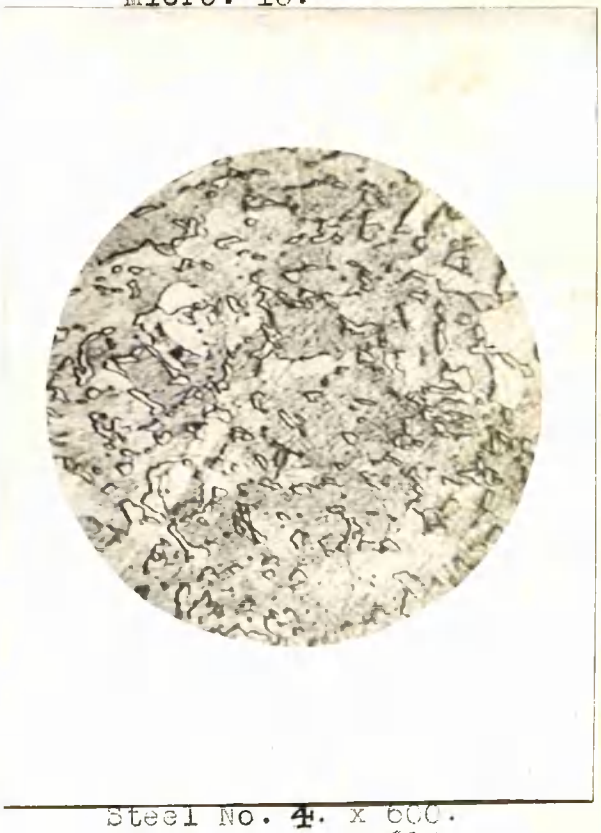
Steel No.4. x 600.
12 Hours at 700°C.

All Etched in 1% HNO_3 in Alcohol.

Micro. 9.

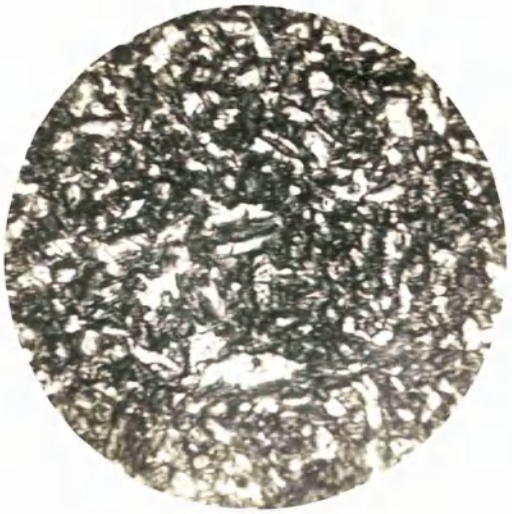
Steel No. 4. x 600.
12 Hours at 705°C.

Micro. 10.



Steel No. 4. x 600.
24 Hours at 705°C.

Micro. 11.



Steel No. 6. x 600.
12 Hours at 700°C.

Micro. 12.

Steel No. 6. x 600.
12 Hours at 705°C.

All Etched in 1% HNO_3 in Alcohol.

Micro. 13.

Steel No 6. x600
24 Hours at 705°C.

Micro. 14.

Steel No. 6. x600.
12 Hours at 710°C.

Micro. 15.

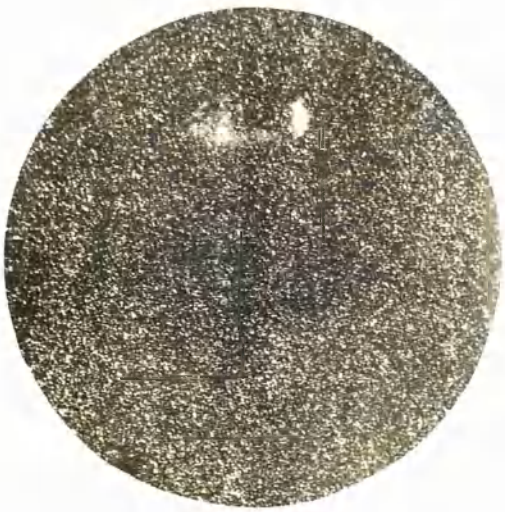
Steel No. 2. x 100.
12 Hours at 700°C.

Micro. 16.

Steel No. 4. x 100.
12 Hours at 700°C.

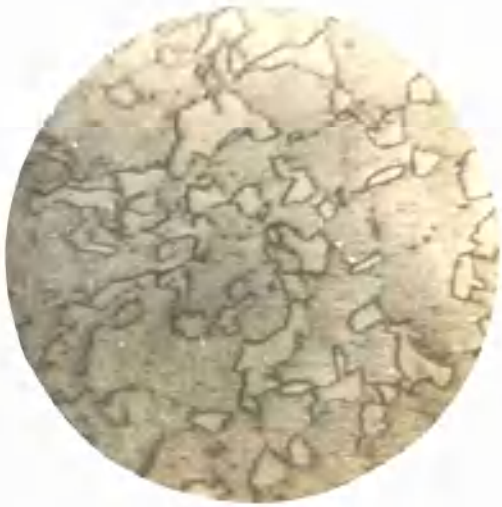
All Etched 1% HNO_3 in Alcohol.

Micro No. 17.



Steel No. 6. x 100.
12 Hours at 700°C.

Micro. 18.



Steel No. 4. 12 Hours 690°C, water quenched.
x 800.